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THE METALLURGY
OF STEEL CASTINGS

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The Metallurgy of

STEEL CASTINGS

BY

CHARLES WILLERS BRIGGS

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VI

PREFACE

This book has been prepared in response to the need in the steel casting industry for a single source reference, complete with detailed information on technical and metallurgical control in the production of steel castings. The book, however, has not been prepared solely for the technical man in the steel casting industry. Rather, it was the purpose to discuss the subjects in such a manner that they would be understood by operating men, workers, and apprentices in the industry, by buyers of castings, design engineers, and metallurgists of other industries, and by students who may wish to know more about the metallurgical and manufacturing details of the steel casting industry.

The point of view taken in this book is one of technical control of all manufacturing operations for the production of steel castings. The quality control of a product cannot be properly formulated without a thorough understanding of the technical problems encountered. It is the sincere hope of the author that this book will provide the foundation for that understanding.

Many of the discussions presented here have been given in lecture form before meetings of technical and operating men of the Steel Founders' Society. The response from these men proved most gratifying, indicating that in addition to the key operating executives a large group of supervisory employees are keenly interested in the technical problems connected with the production of steel castings. For the benefit of these men it has been the plan of the writer to make available a reference work as an extended version of the lectures.

The author is grateful to Dr. R. F. Mehl of the Carnegie Institute of Technology for his assistance with the manuscript. Profound gratitude is also due colleagues in the steel casting industry for their innumerable suggestions and assistance: G. S. Baldwin, Standard Steel Works; H. Blosjo, Minneapolis Electric Steel Castings Company; J. B. Caine, Sawbrook Steel Castings Company; J. Erler, Farrel-Birmingham Company; W. Finster, Reading Steel Casting Division, American Chain and Cable Company; R. A. Gezelius, General Steel Castings Corporation; A. W. Gregg, Whiting Corporation; J. H. Hall, consultant; W. W. Heimberger, Buckeye Steel Castings Company; C. B. Jenni, General Steel Castings Corporation; J. W. Juppenlatz, Lebanon Steel Foundry; F. A. Melmoth; H. D. Phillips and W. J. Phillips, Phillip Foundry Company C. W. Roberts and E. C. Troy, Dodge Steel Castings Company; H. F.

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<div style="text-align: right">CHARLES WILLERS BRIGGS.</div>

CONTENTS

THE METALLURGY OF STEEL CASTINGS

CHAPTER I

BASIC PRACTICE FOR THE PRODUCTION OF STEEL FOR STEEL CASTINGS

BASIC OPEN-HEARTH

Steel melting for castings, as currently carried on in different furnaces, is discussed in Chaps. I and II. The reason these melting methods are given in some detail is that they differ from those used in the manufacture of wrought steel. For instance, the basic open-hearth practice is entirely a cold-melt operation of short duration following melting. Characteristic changes are made in basic-electric melting for castings to ensure a steel of good fluidity. The acid-electric and acid side-blow converter practices are not employed in the manufacture of wrought steel, while acid open-hearth steelmaking is primarily used for castings. For detailed information regarding the construction, operation, and maintenance of open-hearth and electric furnaces and the care and installation of refractories, it is suggested that reference be made to recognized publications on the subjects.

In the steel casting industry in this country, there are approximately 60 open-hearth furnaces being operated by the basic practice, producing a normal rated capacity of about 20,000 tons per month. Under periods of excessive production, this figure increases to 35,000 to 40,000 tons per month. The furnaces vary in size from 10 to 50 tons rated capacity, and all are fired by oil, natural gas, or producer gas. The preferred size of furnace is one having a capacity of 30 tons, with 24 furnaces of this size now operating.

Basic open-hearth steel for castings has been very largely limited to railroad castings, which include bolsters, side frames, couplers, yokes, draft-gear assemblies, snubbers, hopper doors and frames, locomotive beds, truck and driving wheel centers, and tender beds. A large percentage of basic open-hearth steel is also used to produce turbine castings, castings for water-power developments, and large ship castings, such as sternposts, rudder frames, and struts.

1

The basic practice is one that lends itself fairly well to standardization of heats. This is especially true if the scrap charges can be prepared that are similar in content day after day. The process is favored by many producers because its use makes it possible to reduce the phosphorus content and, to a limited extent, the sulphur content during the making of a heat of steel. The common grade of pig iron used contains 0.25 per cent phosphorus. Heats that are comprised of approximately 30 per cent pig iron would require dephosphorization from 0.09 per cent phosphorus to values of 0.03 per cent or less.

Furnace Construction and Lining.—Open-hearth furnaces are usually of the regenerative type, in which the hot gases pass through the checker chamber and transfer their heat to the brickwork. At periodic intervals of 15 to 20 min., the flow of gases through the furnace is reversed, manually or semiautomatically, allowing cold air to enter the hot chambers and heating it to temperatures of about 2000°F. The regenerative principle promotes fuel economy and temperature control; this is the principle that makes open-hearth furnaces practicable.

The roofs of most open hearths, in the capacities used in the foundry industry, are about 12 to 15 in. thick and consist of hard fired silica brick. These brick should have an SiO_2 content of 95 per cent or more, with a CaO content of less than 3.5 per cent. A high fusion point of at least 3100°F. should be maintained, since molten-steel temperatures are but a little lower than those at which brick fuses. No silica brick should be over a specific gravity of 2.38, and the average of a shipment should be less than 2.36.

Expansion allowances of about 1.25 per cent should be made by providing for a large number of small gaps. This can be done by making use of $\frac{1}{8}$-in. felt pads or tar-roofing paper, placed between every other brick in planes perpendicular to the length of the roof and 12 in. apart.

The working face of the roof brick will probably be at temperatures from 3000 to 3050°F. The part of the brick that is above 2675°F. or thereabouts changes to a network of cristobalite crystals. Zone gradations from cristobalite through tridymite back to the unchanged brick structure can be found in the silica roof brick after use in the open-hearth furnace. These zones are responsible for old roofs' spalling along one or the other of these interfaces when they are heated too quickly.

The greatest furnace production achieved without damage to the roof can best be obtained by recording roof temperatures and not allowing the temperature to go above the maximum set temperature. The maximum temperature employed varies somewhat, but in general a temperature range of 3000 to 3050°F. is used. If the roof is given proper treatment, the wear will not be excessive and will be due to a combination of erosion and corrosion by iron oxide and lime. Spalling can be kept to a minimum

if a proper rate of heating is employed, both with a new roof and on reheating an old roof after cooling.

The walls of the furnaces have traditionally been of silica brick, generally about 12 to 15 in. thick, but the development of various improved basic refractories of high thermal-shock resistance has led to the replacement of silica brick in some furnaces. Good results have been obtained by the use of the metal-case-magnesite brick or chrome-magnesite brick. Silica front and back walls in general have shorter lives than do the roofs. Deterioration is caused by absorption of iron oxide from the gas stream, by the cutting action of the flame, by undercutting at the slag line, and by spalling.

Front walls generally are constructed of silica brick. It is reported that some furnaces use chrome or chrome-magnesite brick, unburned chrome, metal-case magnesite, and unburned magnesite. In some cases the bottom four or five courses are magnesite and the rest silica. Average life of silica brick walls is 100 to 150 heats. Basic and neutral brick give about twice the life.

In some cases the back walls are built with silica brick, usually with one course of chrome brick between them and the magnesite hearth. The present trend is to use the basic or neutral brick. Back walls of basic or neutral brick ordinarily last the life of a furnace campaign. With the sloping-back-wall type of construction, almost any refractory brick can be used since it is protected by dolomite or magnesite granulated material, but a basic brick is preferred.

For most installations the basic or neutral bricks are more economical than silica bricks for back-wall, front-wall, and port construction,[1] since (1) silica bricks must in any case have basic bricks as a base; (2) repairs are less frequent; (3) the strain on the roof, due to frequent back-wall repairs, is eliminated; (4) there is less cut in the banks due to silica drip; and (5) higher working temperatures are possible.

The number of heats that can be obtained from the furnace walls before relining vary, depending on a number of factors such as furnace size and construction, steady or intermittent operation, etc. One foundry reports 600 to 700 heats for the roof lining of a 30-ton furnace, while another operator reports 350 heats for a 50-ton furnace. A foundry with a 30-ton furnace states that only 150 heats can be obtained for the front wall, whereas the back wall will stand 600 to 700 heats before relining. Another foundry, also operating a 30-ton furnace, claims 600 heats for both front and back walls. An operator of a 50-ton furnace reports 150 to 200 heats per lining. The life of the furnace ends is usually similar to that recorded for the front wall.

Bottom Construction.—Basic open-hearth bottoms consist of fire-clay brick laid on the pan. A course of chrome brick is placed above this, and

on top of this there is then placed a course or two of basic brick.   In some foundries the basic brick layers are omitted.   On top of the brick is laid a fused layer of dead-burned magnesite mixed with 15 to 25 per cent open-hearth slag.   In a couple of foundries magnesite brick are used for the entire bottom.   The modern trend, however, is to replace the use of magnesite brick with fire-clay brick capped with chrome or chrome-magnesite brick.

During the past few years insulation material has been employed in a few furnaces between the metal shell and the fire-clay brick.   A typical open-hearth bottom of today may have $2\frac{1}{2}$ to 4 in. of insulation, $4\frac{1}{2}$ to 13 in. of fire-clay brick, 6 to 12 in. of magnesite, chrome, or chrome-magnesite brick, and 6 to 12 in. of sintered magnesite plus slag mixtures.

A recent development is the use of dead-burned magnesium oxide clinker for bottom material.   This material is graded to size and bonded to yield on ramming a dense, strong structure.   This material is laid down on the brick bottom and rammed to the proper shape.   In some cases it has been thought necessary to place a layer of sintered magnesite on top of this, but in two cases in the foundry industry the rammed magnesium oxide clinker forms the top surface of the bottom construction.   The depth of the material varies, depending on furnace construction and size.   Depths from 6 to 15 in. are being used.

After the bottom is rammed in, the furnace is dried with gas and wood fires.   Oil is then used to raise the temperature to 2900°F.   This temperature is held until the moisture stops seeping through the pan.   The hearth is then slagged, the slag tapped, and the furnace allowed to cool to the range 1500 to 2000°F.   The furnace is then brought back to heat, and the cracks in the bottom are covered with fine sintered magnesium oxide material.   It has been estimated that from 4 to 6 days can be saved by installing the rammed bottom over the time required for the preparation of a good sintered magnesite bottom.

If a sintered bottom of dead-burned magnesite is required, the plan usually followed in the steel foundry industry is to mix three parts of magnesite and one part of pulverized basic open-hearth slag.   This material is spread over the furnace bottom as evenly as possible in layers about $\frac{1}{4}$ in. thick.   Time is allowed for each layer to sinter properly before the next layer is placed in the furnace.   One foundry has a rule that, for the first 3 in. of bottom placed in the furnace, an interval of not less than $2\frac{1}{2}$ hr. should elapse as each $\frac{1}{4}$-in. layer is placed in the furnace.   The time necessary to build such a bottom of 12 to 18 in. in a 30-ton furnace is about 6 days.   A sintered bottom can be constructed in a much shorter time of approximately 90 hr. by the use of clinker dolomite.   One foundry reports that for the past 18 years they have been using clinker dolomite for the construction of new bottoms.

The taphole is prepared by laying a 6-in. pipe, coated with basic material, in its proper position and building up the bottom around it. Other methods, such as the use of a wooden block covered with the basic insulating material, are used.

After the sintered bottom is put in, pulverized slag is placed on the banks and allowed to melt and run down into the bottom. This is repeated at intervals and continued until there is a depth of 3 or 4 in. of liquid slag on the furnace bottom. The time allowed for slagging a bottom of a 30-ton furnace is about 6 hr. The slag is then tapped, and the furnace is ready for the first heat.

Fuels.—There are two types of fuel used for basic open-hearth furnaces in the steel foundry: oil and gas (natural and producer). The fuel oil used varies in gravity from 5 to 20°Bé. at 60°F., depending on local furnace conditions. Some foundries prefer a gravity of about 15, while others prefer a 10°Bé. oil. Fuel oils normally used have a B.t.u. value per gallon of approximately 150,000. The oil is heated from 150 to 200°F. before it enters the burner. Various types of oil preheaters are used, but those employing steam for preheating are the most popular.

Nearly all foundry open-hearth operators place a sulphur limitation on the oil used, and the majority of operators have selected 0.75 per cent sulphur as a maximum specification. If the oil contains sulphur contents under this value, sulphur pickup of the bath will not constitute a serious factor.

The fuel oil is injected into the furnace by the use of either air or steam. Some operators are equipped to use both air and steam, such as air in the summer and steam during the winter. The temperature of the superheated steam used for atomizing the oil is 600 to 900°F. at the regulator; and a pressure of from 50 to 125 lb. is employed, depending on the type of oil burner. This pressure is varied from time to time according to furnace conditions, but it is usually held at the top value as established by each foundry until the heat is melted.

Some furnaces in the foundry industry are equipped with means of using both oil and gas. In some cases one acts as an auxiliary to the other, while in other cases one is used to hold the furnace temperatures during a week-end layover and the other is used for melting purposes. Natural-gas pressure at the burner will vary, depending on the type of gas burner employed. The low-pressure burners operate on gas pressures from 1 to 5 lb., and 20 to 35 lb. gas pressure is used in the high-pressure burners. The B.t.u. value of the gas usually runs between 950 and 1000 B.t.u. per cu. ft. Producer gas requires regeneration, while natural gas may be burned direct. In the case of producer gas, the pressure is usually 1 to 2 in. water, and large burners or ports are used. The B.t.u. value is low, ranging from 150 to 160.

The amount of fuel being burned per hour is indicated and recorded. For a 30-ton furnace, the fuel consumption will be from 28 to 35 gal. per ton of metal charge, based on normal operations over a period of 30 days. About 25 gal. per ton will constitute the fuel consumption per ton charged during the period of a heat cycle (from tap to tap). With producer gas firing, the fuel consumption is in terms of coal. Ordinarily, 550 to 600 lb. of gas coal are required per net ton of steel melted.

The type of fuel used will make some difference in the oxidation of the steel. A gas-air flame is a "soft" flame, which rolls easily through the furnace. If low-pressure gas is used, all the air is supplied by the forced-draft system, whereas if high-pressure gas firing is maintained, a pressure of only about 35 p.s.i. is employed. The oil-air flame is a "hard" flame, which is forced through the furnace. Some of the air is furnished by a forced-draft system; but high-pressure air is required, approximately 100 p.s.i. to atomize the oil. This hard flame rushing through the furnace carries an excess of oxygen with it and causes greater oxidation than a soft gas-air flame. This often means that the proper degree of oxidation may be obtained in oil-fired furnaces with less iron ore than would be required in a gas-fired furnace. Also, if no ore were used in either type of fuel-fired furnace, the one fired by oil would have a faster oxidizing period.

The Charge.—The charge of the basic open-hearth furnace consists of pig iron, purchased scrap, foundry returns (gates, risers, etc.), lime, and ore.

Pig Iron.—The pig iron should contain a high manganese content of 1.50 to 2.00 per cent, in order that full advantage may be taken of its desulphurizing action. While it is not necessary to require a minimum of 1.50 to 2.00 per cent manganese in the pig iron when making the carbon steels, it is generally agreed that to do so results in better steel, especially in its impact properties, by virtue of the lower sulphur content. There is also an advantage in the better working of the heat with the higher manganese content.

The silicon content of the pig iron should be kept low, probably 1.25 per cent maximum, to prevent undue corrosion of the basic-furnace banks by the acid SiO_2. Likewise, there is less possibility that the oxidized and rejected phosphorus may leave the slag and reenter the steel bath, since a high degree of basicity must be maintained in the slag to enable it to retain phosphorus. High silicon in the bath will produce a thin slag, the correction of which will require increased lime additions and will result in a large slag volume.

The sulphur content of the pig iron should be as low as possible. It may contain a maximum of 0.06 per cent, provided the fuel is low in sulphur and the iron high in manganese, although even this value is too high

for some foundries. The phosphorus content should be less than 0.35 per cent.

In the purchase of scrap for the basic open hearth, selection should be made on the basis of analysis and physical condition. Sampling and analysis should be carried on to foretell the presence or absence of certain elements, such as sulphur, and other residuals, as chromium, copper, and molybdenum, which may affect the ready weldability of carbon steels.

Lime.—The principal raw-material constituent of basic slag is lime or limestone. This limestone should be uniform in composition, low in MgO and SiO$_2$, and high in CaO contents. In nearly all basic open-hearth steel foundries, burned lime is used in place of limestone. The reasons for this are that the time for melting is less because of the fact that it is not necessary to absorb heat in the calcination of the limestone, a more uniform CaO content can be maintained by its use, and a good slag covering is formed early in the heat.

Lime used in place of limestone usually is classified as metallurgical lime and is lump size or is graded from "rice size" to "pebble size" with all dust screened out. The rice size is usually supplied in airtight steel drums to prevent it from slacking. Lump- or pebble-size lime is shipped in box or container cars. Specifications for metallurgical lime are as follows:

	Per Cent	
Loss on ignition	2	max.
Silica	1	max.
Iron oxide and alumina	1.5	max.
Magnesium oxide	1.5	max.
Calcium oxide	95	min.

Some foundries make a practice of adding manganese ore with and in place of a percentage of the iron ore. This practice increases the MnO content in the slag and, it is claimed, assists in the working of the heat and in the sulphur removal.

Charging.—The furnace is nearly always charged by covering the bottom with light scrap such as rails when available, plate scrap, or structural or scrap car sides. This acts as a bed for the lime additions and prevents the lime from sticking to the furnace bottom. Also, the light scrap covering charge protects the bottom when the heavy material, such as large risers, is charged. The lime should be spread over the scrap as evenly as possible and should be followed by the balance of the scrap and the foundry returns. The pig iron usually is placed on top of the charge. Because of the bulkiness of the scrap, it may be necessary to charge it in two or three additions.

The make-up of the charge for the average carbon heat varies somewhat, depending on foundry conditions and practice. Some foundries

have a higher foundry scrap return than others because of the class of work done. Others, on account of the location, find it more profitable to use a high pig-iron charge, finding it difficult to obtain the type of scrap that they desire. Charges at four different plants are as follows:

	Foundry A	Foundry B	Foundry C	Foundry D
Pig iron.................	10	25	30	25
Own scrap...............	15	10 .	50	40
Purchased scrap..........	75*	65	20	35

* Includes high-carbon springs and cast-iron motor blocks.

Once a type of charge is chosen, every effort is taken to see that the proportions of the charge are followed in successive heats. The purchased scrap may be classified into types, and definite percentages of each are fixed so that the following heat will be as nearly like the preceding one as possible.

Charges are made up to a definite carbon content and, in some cases, to a definite manganese content. The carbon and manganese contents selected will be governed by the working conditions of the furnace. Some furnaces permit greater oxidation during the melting period than others.

Under normal conditions the amount of lime charged will vary from 4.5 to 6 per cent of the weight of the metal charge. The lack of selectivity of scrap during the Second World War forced many melters to increase the percentage of lime in the charge, and in some heats the lime content was increased to 7 per cent of the charge. The figure used depends upon the slag basicity ratio desired. If limestone is used, it will constitute about 10 per cent of the metal charge.

In some cases, petroleum coke or other carbonaceous materials will be substituted in part for the pig-iron content. This procedure is carried on when pig iron is difficult to obtain or the price is high. The coke supplies the carbon equivalent of that found in the pig iron, so that a fairly definite carbon content will be obtained in the metal at the time it is completely melted.

The carbon content of the charge will vary somewhat, depending on the carbon content desired at melt-down. In general, the carbon content of the charge will be from 1.00 to 1.75 per cent. A fast-melting furnace will not require the charge to be as high in carbon as a slower working furnace. Also, the type of the scrap will be influenced in determining the carbon content of the charge. The use of limestone in preference to lime requires, usually, a higher carbon content in the charge in order to arrive at similar carbon contents at melt-down, owing to the loss of

carbon from the metal during the calcination of the limestone (*e.g.,* during the lime boil).

The silicon content of the charge will vary from about 0.50 to 0.85 per cent, depending on the quantity of pig iron used in the charge. Little consideration is given to the maintaining of a fixed silicon content in the charge. The manganese content of the charge also varies under rather wide limits from about 0.45 to 1.25 per cent, depending mainly on the amount of pig iron used in the charge. If considerable amounts of purchased scrap are used, then the manganese content of the charge may be rather low as rolling-mill scrap will average only about 0.30 per cent manganese.

Special consideration was given during the Second World War to the possible increase of production of a furnace by the use of cupolas to furnish hot metal in place of the cold pig charge conventionally used. It is too early to give conclusive reasons for its use, but those who advocate such a practice outline the following advantages:

1. Small-size, lightweight charging scrap can be readily handled through the cupola to furnish liquid iron to form an easily handled part of the charge. The result of liquid iron can readily be held to a maximum of 3 per cent carbon and to almost any silicon content.

2. When pig iron is obtainable but is of a higher silicon content than can safely be used, it can be melted in the cupola and can be adjusted to the desired silicon content by (*a*) incorporating low-silicon steel scrap with the pig iron charged in the conventional manner; (*b*) charging all high-silicon pig and increasing the air blast in the conventional cupola, thus oxidizing a good part of the silicon; (*c*) charging all high-silicon pig and raising the tuyères of the conventional cupola so that a good part of the silicon can be oxidized.

3. High-sulphur steel scrap or pig can be used, since a ladle treatment with alkali will readily lower the sulphur to a maximum of 0.06 per cent.

4. A probable saving of 30 per cent production time results, since as much as 30 to 40 per cent of the charge is in the molten condition.

The Melt.—The furnace is operated with a full sharp flame during the melting period, predetermined settings being maintained by instruments. As the charge melts, slag is formed, because of oxidation to SiO_2 and Fe to oxide. A bubbling action may be observed on the bath during the melting-down period; this is due to the release of CO_2 gas on the calcination of limestone, if limestone is used, and is known as the "lime boil." If lime is used, it will come up through the bath in partially melted lumps. Some of the lime may stick to the bottom and not come up until the charge is melted and the bath fairly hot. Those interested in production speed may assist in getting the lime into solution by adding ore to the bath before it is completely melted.

As soon as the heat is melted and the lime is up, which is indicated by the absence of localized violent boiling action in the bath, a sample of the bath should be taken.   Also, at this time an adjustment to a softer flame is made, although some operators use the sharp flame throughout the heat.

The carbon content reported on the first analysis will give the melter an indication of the amount of iron ore needed and will foretell whether or not the selected charge melted in accordance with previous heats. Steel-casting practice requires that at the time the bath is completely melted, the carbon content is generally from 0.25 to 0.40 per cent above the finishing carbon content.   Three open-hearth requirements of melt-down carbon for a 0.25 to 0.30 plain carbon steel are as follows:

Plant	Carbon, Per Cent
A	0.50–0.60
B	0.60–0.70
C	0.50–0.55

It is believed that the best rate of production is obtained on heats melting at 35 to 50 points of carbon above the desired tapping carbon, whatever this may be.

During the oxidation period the carbon will drop to about 0.20 per cent, and the finishing additions will bring the final carbon back to the 0.25–0.30 per cent figure.   The loss of 40 points of carbon in approximately 1 hr. gives to the bath a very vigorous boil or bubbling action. The silicon content at melt-down is at a low figure, seldom over 0.02 per cent and usually even less.   The manganese content at melt-down will vary from about 0.20 to 0.35 per cent, and it will depend on the total manganese content of the charge, the melting speed of the furnace, and MnO content of the slag.   Many operators like to keep the manganese-residual content high, and a very good way of doing this is to have a manganese content of about 0.35 per cent at melt-down.

The Oxidation Period.—Iron-ore additions are made to the bath as soon as it is melted and has reached a temperature in excess of about 2700°F.   The ore addition is made in sufficient quantities to produce a long, vigorous boiling action.   The iron ore added to the bath oxidizes the carbon in the melt, producing CO.   The amount of iron ore added will depend on the value of the melt-down carbon and on the finishing carbon desired and will usually be about 0.50 to 1.00 per cent of the metal charge.   A good, vigorous boil is desired, even though it is necessary to go below the carbon content desired, and thereby require the added operation of recarburizing.   Some operators deliberately plan the boil so that it is necessary to recarburize.   It is their thought that the steel has a greater degree of fluidity and less chance of included gases such as

hydrogen and nitrogen. Differentiation should be made between the violent localized boiling due to limestone, scrap, and additions and the over-all boil or working at the slag-metal face. The violent local boils are effective for gas release and can be obtained by deeply stirring the bath with clean steel bars.

The boil should be vigorous during the first part of the oxiuation period and should taper off during the latter part with the rate of carbon drop being substantially higher for the initial period of approximately $\frac{1}{2}$ hr. As the carbon drops, the boiling action becomes uniformly gentle, and at the end of the oxidation period, when the carbon content is in the neighborhood of 0.20 to 0.25 per cent, the rate of carbon drop is much reduced. The rate of carbon drop will vary somewhat, depending on the working speed of the furnace and construction of the bath. Rates of carbon drop for three furnaces are shown in Table I. The amount of fuel, air, and draft is usually gradually diminished during the oxidation period. A positive pressure is, however, maintained.

During the period of carbon oxidation the manganese is partially oxidized and passes into the slag. The distribution between charge and slag is in a definite ratio and is determined by the temperature and the composition of the slag. If the total manganese present is high, the charge will contain a high content of manganese, and the slag a high proportion of manganese oxide. Thus, if manganese ore is added along with the iron ore, the MnO content of the slag will build up to maintain a high-manganese residual. Also, for a given manganese charge (all sources), the higher the bath temperature, the higher will be the residual manganese. A high residual manganese will permit a high yield of alloys at the time of deoxidation.

The presence of manganese oxide in the slag reduces the content of ferrous oxide in the slag and hence the oxygen supplied to the charge. These considerations have led some operators to maintain a high concentration of manganese by introducing manganese to the charge at the outset. However, manganese apparently plays an important part in the reaction only toward the end of the oxidation period, since before this point is reached, the carbon content of the charge, owing to its high reducing action, alone governs the oxygen content of the charge. In addition, at the start after the charge has been melted and the temperature is still comparatively low, there must be a supply of ferrous oxide from the slag to allow the charge to boil vigorously. If the manganese in the charge is high, a good boil can only be obtained by adding large quantities of ore, which unnecessarily increases the quantity of slag. For this reason, it has been the practice of some operators to increase the manganese content of the bath and the slag by an addition of spiegel toward the end of the oxidation period.

Phosphorus can be eliminated from the metal bath during the oxidation period; in fact, the removal of phosphorus depends on the existence of oxidizing conditions, which in turn are controlled by the FeO/MnO ratio in the slag. The use of a high-manganese charge (high MnO slag) tends to retard dephosphorization of the bath by diminishing the state

TABLE I.—RATE OF CARBON DROP FOR VARIOUS BASIC OPEN-HEARTH FURNACES

Time, after ore addition, min.	Plant A, 25-ton		Plant B, 50-ton		Plant C, 30-ton	
	Carbon content, per cent	Rate of C drop per 10 min., per cent	Carbon content, per cent	Rate of C drop per 10 min., per cent	Carbon content, per cent	Rate of C drop per 10 min., per cent
0	0.58		0.51		0.60	
10	0.49	0.09	0.43	0.08	0.50	0.10
20	0.40	0.09	0.38	0.05	0.44	0.06
30	0.33	0.07	0.33	0.05	0.36	0.08
40	0.26	0.07	0.27	0.06	0.29	0.07
50	0.21	0.05	0.23	0.04	0.25	0.05
60			0.18	0.05		

TABLE II.—COMPOSITION OF ALLOYS

Name	Per cent								
	C	Mn	Si	Cr	Mo	Ca	V	Ti	Al
Spiegeleisen (spiegel)...............	4.5–5.0	15–30	1.0						
Silicomanganese No. 4.............	Max. 3.0	65–70	12–14						
80% ferromanganese (standard).....	6.0–8.0	78–82	Max. 1.0						
50% ferrosilicon (standard).........	Max. 1.0		47–52						
15% ferrosilicon....................	Max. 1.0		14–16						
Ferrochrome.......................	4.0–8.0		2–3	60–72					
Ferromolybdenum.................	Max. 2.0		Max. 1.50		55–65				
Ferrovanadium....................	0.2–3		1.5–8				35–46		
Calcium silicon....................			60–65			28–35			
Calcium manganese silicon.........		8–10	55–60			17–19			
Ferrotitanium, (high-carbon).......	7.5		2.5					17	1.0
Alsifier............................			40.0						20.0

of oxidation. Thus a slag having a total FeO content of 18 per cent and a MnO content of 11 per cent would be more effective in the removal of phosphorus than one with an FeO content of 12 per cent and a MnO content of 11 per cent, provided that the basicity of the slag is the same in both cases.

The presence of slags high in lime assists in the phosphorus removal. The action of lime is only secondary, for in the absence of the FeO the

lime will not by itself remove phosphorus from the steel. A lime content of 40 to 45 per cent is desirable, or a lime-silica ratio of about 3.0 to 1 should be maintained for efficient phosphorus removal.

Some foundries fix the time that must elapse after the last ore addition and before the metal is tapped from the furnace. This time period varies from 40 to 90 min., depending on the size and working conditions of the furnace, such as the slag and the carbon content of the bath, and the desired finishing carbon. Ore should have sufficient time to work through the bath and to produce a uniform and slow carbon drop. One operator proceeds on the basis that no ore is added after the carbon content has dropped to 15 points above the desired tapping carbon, which in his practice is not usually lower than 0.45 per cent carbon.

Refining.—The refining period is actually the last part of the oxidation period. During this period, no ore additions are made and the carbon drop is uniform and slow. The temperature of the bath is fairly high, and the slag conditions are properly adjusted to produce the final slag.

During this period, the high temperature assists in the sulphur removal, provided that the slag has a high CaO content and the active FeO content is low. If the FeO/MnO ratio is low, the slag will not be so oxidizing in character and there will be more chance of sulphur elimination. It is at this time that spiegel (Table II) is added by some operators. The manganese content of the bath is increased, which also aids in the sulphur elimination, the MnO content of the slag is increased, and this in turn lowers the FeO/MnO ratio. The addition of the spiegel to the bath acts as a partial "block," reducing the oxidation conditions of the bath, and is effective in bringing about a slower drop in the carbon content. Additions of fluorspar are made to the bath at this time by some operators in order to keep the slag in a fluid condition, while further additions of lime are made to the bath in order to increase the CaO content of the slag.

Studies have been made of the inclusion count of basic open-hearth steel in relation to the FeO content of the slag just prior to blocking. Regan[6] reports the following:

Total FeO, Per Cent	Average Per Cent of Inclusion by Inclusion Count
10.4	0.009
15.5	0.0155

The higher the iron oxide content, the more inclusions present in the cast steel.

Blocking and Deoxidation.—When the carbon content has dropped to a point in keeping with the desired final analysis, the further drop in the carbon content is stopped by the addition of deoxidizing alloys, which

combine with the oxygen, removing it from reaction with carbon. This operation is known as "blocking the bath," and the ferro alloys of manganese and silicon are used for this purpose.

The alloys used for the block vary slightly from plant to plant; *e.g.*, one foundry uses a spiegeleisen and silicomanganese block. Spiegel of about 4 per cent silicon is used in quantities of about 40 lb. per ton. Another foundry has a preference for 15 per cent ferrosilicon, while still another uses spiegeleisen and ferrosilicon as the block.

The use of a 15 per cent ferrosilicon block 15 to 20 min. before tapping is practiced by several foundries. Such a practice, however, has a tendency of driving phosphorus back into the metal from the slag. Since reversion of phosphorus to the bath is certainly not desired, it has been suggested that the use of the silicon block be discontinued. If operations are set up in order that careful checking of the carbon and manganese contents of the bath can be made by rapid analytical methods, the use of the block is unnecessary. The elimination of the silicon block results not only in a lower phosphorus content but in less opportunity for the absorption of gases, such as hydrogen and nitrogen.

One foundry reports that they seldom block heats with any alloy of silicon. Instead, the slag is thickened with additions of burned dolomite, and the bath is allowed to remain active until the time of the ferromanganese addition. This method gives good manganese recovery, lower sulphur contents, and little opportunity for phosphorus reversion. All silicon is added to the ladle, thereby affording little opportunity for gas pickup.

Tapping.—After the blocking alloys are added, a period of 8 to 10 min. is permitted to elapse before the heat is tapped to allow for the alloys to melt and be well distributed throughout the bath. When the final additions are made, the draft, the steam or air used for atomizing, and the amount of fuel oil are again reduced to a low value and kept at this point until the heat is tapped. The heat is tapped by thrusting the tapping rod through the bath and punching out the bottom material in the taphole.

Ladle Deoxidation.—When the metal in the ladle is approximately 1 ft. in depth, the 50 per cent ferrosilicon is added to the stream of metal as it runs from the furnace, and the 80 per cent ferromanganese is added directly to the ladle. In some cases the reverse procedure is used.

The manganese recovery will vary from 65 to 95 per cent, depending on the deoxidation practice, the residual manganese, the MnO content of the slag, and the FeO/MnO ratio. If ferrosilicon is used as the block, or if ferrosilicon is added prior to ferromanganese, the manganese recovery will be higher than if ferromanganese or spiegeleisen is used first. A higher residual manganese, a high MnO content in the slag, and a low

FeO/MnO ratio are all responsible for a high-manganese recovery.   One foundry reports that with 80 per cent ferromanganese added to the furnace on heats of steel to contain less than 0.90 per cent manganese, a recovery is obtained of three-quarters of the manganese in steels of 0.30 to 0.40 per cent carbon and two-thirds of the manganese in steels with a carbon content of 0.10 to 0.20 per cent.   If 80 per cent ferromanganese additions are made to the furnace when producing high-carbon steels, 1.00 per cent or higher, the recovery drops sharply.   Additions to the ladle seem to be fairly consistent at 80 per cent recovery under normal conditions.

The silicon recovery will be between 90 and 95 per cent if all is added to the ladle and the silicon block is not used.

The present trend in alloy practice is to add the so-called "exothermic" alloys such as Chrom-X and Sil-X to the ladle.   Chromium recovery is approximately 80 per cent.   This is regardless of the percentage of chromium introduced through the scrap charge.

If additional recarburization is necessary, depending on the last reported carbon analysis, to bring the metal within the prescribed limits set, then a carbonaceous material such as anthracite coal or petroleum coke is added to the ladle or to the stream of metal as it leaves the furnace.   Additions of coal or coke are made at the rate of about 0.25 to 0.30 lb. per ton per 0.01 per cent carbon increase.

It is only occasionally necessary that an addition of a special deoxidizer, such as an aluminum, is made in the ladle to basic open-hearth heats for steel castings.   Grain-size control is not very often specified, although more grain-size requirements may be specified during the postwar period.   From 2 to $2\frac{1}{2}$ lb. aluminum addition per ton of steel are used to produce fine-grained, medium-carbon steels.   Aluminum shot is an efficient method for effecting good recovery.   In some cases aluminum is used as a special deoxidizer when the heat is scheduled for green sand molds.   However, in this case the aluminum is more often added directly into the metal stream as it enters the mold.

In some cases additions of special alloys are added to increase hardenability.   Additions of ferroboron, or a combination alloy containing boron, are made in the ladle after all other deoxidizers have been added. The aluminum content present should be sufficient to produce a fine-grained steel.   Boron equivalent to 0.003 per cent is added with a recovery of about 0.0025 per cent expected.

The tapping temperatures will range from about 2900 to 3000°F., depending on the size of the molds and the number of molds that are to be poured.   The temperature is judged by the aid of film or set spoon tests.   Each type of steel will form a film at various times, depending on analysis and temperature.   These times are correlated and goals are set

according to the number of castings and type of steel.   Another procedure used is to judge temperature by pouring steel from a spoon test.

The time required to melt a heat will of course depend on the size of the heat.   A 30-ton charge can be melted in about 4 hrs., and the time required for the total operation is about 5 or $5\frac{1}{4}$ hr.   A review of furnace operations in steel foundries reveals that the time the total operation takes in minutes per ton of steel charged varies from 9 to 12 min., with an average of about 10.5 min. per ton.   The larger furnaces usually take the shorter time in minutes per ton of steel melted.

A log of a typical basic open-hearth heat is given in Table III.   A log showing the addition of spiegeleisen to increase the MnO content is shown in Table IV, and in Table V are presented a log and test analysis of a heat prepared with the use of manganese ore to maintain a high MnO slag and high-manganese residuals throughout the heat.

Slag Control.—The primary aim of slag control is to regulate the rate of reactions and thereby regulate the degree of oxidation of the steel bath. The object of slag control for the basic open-hearth practice applicable to killed steel for castings is as follows:

1. To bring about as quickly as possible the correct carbon analysis.

2. To produce a slag that will eliminate phosphorus and sulphur to the desired extent without excessive use of lime or excessive loss of iron as oxide.

3. To conserve deoxidizers.

4. To standardize the conditions of oxidation so that the deoxidizing additions will unfailingly produce the desired type of steel for castings.

Slag control should result in greater uniformity of heats, time, heat consumption, chemical and mechanical properties, and savings on deoxidizers and slag-forming materials.

It may thus be seen that the degree of slag deoxidation has an important bearing on the oxidation of the steel.   Oxygen control is the most important item in the various steelmaking methods.   In the basic practice, control of phosphorus and sulphur can also be effective through careful slag manipulation.   In order to arrive at a desired composition of the metal with particular respect to oxygen, phosphorus, and sulphur, it is necessary to employ a definite type of slag.

Knowledge of the variables affecting oxygen, sulphur, and phosphorus contents of the metal has been summarized by Herty[3] as follows:

Oxygen.—The oxygen content of the metal before deoxidation is controlled by the carbon in the metal and the FeO content of the slag.   Above about 0.10 per cent carbon, the slag is of minor importance, but as the carbon diminishes below 0.10 per cent, the slag becomes more important.   After deoxidation, particularly in the furnace, the slag composition and fluidity play an important role in determining the oxygen content of the steel.   Temperature increases the solubility of FeO in iron,

TABLE III.—LOG OF A 25-TON BASIC OPEN-HEARTH HEAT

Size of heat: 50,000 lb.

Composition desired: Carbon, 0.25 to 0.30 per cent; manganese, 0.65 to 0.75 per cent; silicon, 0.35 to 0.45 per cent; phosphorus and sulphur, 0.035 per cent max.

Charge:

	Lb.	Per cent
Pig iron	7,500	15
Forgings	9,000	18
Heavy melting	17,500	35
Punchings	8,000	16
Cast steel	8,000	16
	50,000	
Lime	2,500	5

Time	Activity
11:40	Started charging
1:30	Finished charging
3:30	Melted (3 hr. 50 min. melting time)
3:30	First test: carbon, 0.58 per cent
3:35	Iron-ore addition 500 lb. (1.0 per cent)
3:50	Second test: carbon 0.42 per cent; manganese 0.21 per cent
4:00	Third test: carbon 0.36 per cent; manganese 0.20 per cent
4:00	Lime 750 lb.
4:10	Fourth test: carbon, 0.29 per cent
4:15	Fluorspar 75 lb.
4:20	Fifth test: carbon 0.26 per cent; manganese 0.21 per cent
4:23	15 per cent FeSi: 400 lb.
4:33	Tapped (time for heat 4 hr. 53 min.)

80 per cent FeMn addition to ladle.............. 294 lb.

50 per cent FeSi addition to stream.............. 510 lb.

Petroleum coke to ladle........................ 15 lb.

Final analysis: Carbon, 0.28 per cent; manganese, 0.69 per cent; silicon, 0.40 per cent; phosphorus, 0.022 per cent; sulphur, 0.027 per cent.

Slag analysis taken prior to the addition of the 15 per cent FeSi block:

	Per Cent
CaO	46.1
SiO_2	15.3
FeO_5	13.7
$Fe_2O_?$	3.2
MnO	9.1
MgO	6.5
Al_2O_3	3.4
P_2O_5	1.8
CaO/SiO_2	3.0
Total FeO	15.1

TABLE IV.—LOG OF A 50-TON BASIC OPEN-HEARTH HEAT

Size of heat: 100,000 lb.

Analysis desired: C, 0.22–0.27 per cent; Mn, 0.65–0.75 per cent; silicon, 0.30–0.40 per cent; phosphorus and sulphur, below 0.03 per cent.

Charge:

	Lb.	Per cent
Pig iron	25,000	25
Heavy melting	25,000	25
Forgings	10,000	10
Cast-steel returns	40,000	40
	100,000	
Lime	6,000	6
Fluorspar	400	0.4

Time	Activity
	Time required to melt: 7 hr. 55 min.
10:00	Melt-down—first test: carbon 0.68 per cent (FeO 10 per cent, CaO/SiO$_2$ 2.4 per cent)
10:00	Iron-ore addition 800 lb., 0.8 per cent
10:30	Second test: carbon, 0.55 per cent; Mn 0.23 per cent
11:00	Third test: carbon, 0.41 per cent; Mn 0.20 per cent
11:00	Spiegel: 600 lb. (0.6 per cent)
11:30	Fourth test: carbon, 0.30 per cent; Mn 0.25 per cent
12:00	Fifth test: carbon 0.19 per cent; Mn 0.27 per cent
12:13	Sixth test: carbon, 0.15 per cent
12:13	Spiegel (4 per cent Si) 2,000 lb.
12:18	Silicomanganese 400 lb.
12:28	Tapped

50 per cent FeSi, ladle...................... 350 lb.

Anthracite coal to the stream................. 20 lb.

Ladle analysis:

Carbon, 0.25 per cent; Mn, 0.74 per cent; silicon, 0.35 per cent
Phosphorus, 0.02 per cent; sulphur, 0.024 per cent

Melt-down slag:

SiO$_2$	CaO	FeO	MnO	P$_2$O$_5$	CaO/SiO$_2$
20.1	49.5	10.1	10.0	1.41	2.5

Finishing slag:

SiO$_2$	CaO	FeO	MnO	P$_2$O$_5$	CaO/SiO$_2$
17.8	53.2	14.7	7.6	1.74	3.0

TABLE V.—LOG OF A 30-TON BASIC OPEN-HEARTH
Porter[1]

Charge:

Metal scrap.............................	60,000	
Tar coke................................	250	
Burned lime.............................	2,600	(4.3 per cent)
Manganese ore..........................	600	(1.0 per cent)

Refining of heat:

Time	Activity
12:35	Melted down
12:35	Iron ore 400 lb.
12:50	Iron ore 200 lb.
1:14	15 per cent ferrosilicon 500 lb.
1:22	Heat tapped

80 per cent ferromanganese in ladle.............	244 lb.
50 per cent ferrosilicon in runner...............	540 lb.
Tar coke in runner	20 lb.

SLAG ANALYSIS AND PRELIMINARY AND FINAL METAL ANALYSIS

	Melt test.* per cent	Second, per cent	Third, per cent	Fourth, per cent	Fifth, per cent
SiO_2......	14.81	17.61	16.88	16.21	15.58
Al_2O_3...............	2.44	2.48	2.16	2.05	2.01
Fe_2O_3	3.34	3.51	3.72	4.02	4.40
FeO.................	9.62	13.63	15.44	14.39	15.11
MnO...............	12.54	12.02	11.35	11.28	10.81
Cr_2O_3.	0.29	0.37	0.40	0.36	0.35
CaO...............	47.93	42.20	40.94	42.36	43.14
MgO....	6.02	5.57	6.42	6.52	5.77
P_2O_5..	1.96	1.95	1.94	2.22	2.31
S........	0.15	0.19	0.21	0.25	0.28
Time taken....	12:30 P.M.	12:45 P.M.	1:00 P.M.	1:08 P.M	1:13 P.M.

Preliminaries, per cent

Carbon................	0.60	0.40	0.36	0.28	0.25
Manganese.............	0.34	0.33	0.35	0.34	0.35
Silicon.................	0.016	0.016	0.006	0.002	0.002
Phosphorus.......... .	0.012	0.007	0.007	0.007	0.009
Sulphur...............	0.035	0.032	0.031	0.029	0.029

Final analysis, per cent

	C	Mn	Si	P	S
35th mold.............	0.27	0.63	0.40	0.021	0.029

* *Note:* At the same time the metal test was taken, there was also a slag test poured.

affects the carbon–iron oxide reaction in the metal, and increases diffusion of iron oxide from slag to metal.

Phosphorus.—The phosphorus in the metal is lowered by increased basicity and fluidity of the slag, a high FeO content of the slag, and by low temperatures. High phosphorus in the charge is an adverse factor to the production of low-phosphorus metal.

Sulphur.—The sulphur in the metal is lowered by increased basicity, fluidity, and agitation of the slag and by high temperature. High iron oxide and increased sulphur in the charge make the attainment of a given sulphur more difficult.

Estimation of Slag Composition.—In order to control oxygen, phosphorus, and sulphur in the metal, knowledge must be acquired of the carbon content of the metal, the FeO content of the slag, and the basicity of the slag during the heat. The appearance of the slag in the furnace during the working of the heat is indicative to experienced steel melters of its composition. Furthermore, the color of the slag after it has been cooled in water has also been used as a guide to its approximate composition (Table VII). The appearance of the surface texture, color, and fracture of slag cakes poured into standard molds is a useful guide to slag composition.[13]

Slag and Metal Analysis.—Rapid methods of chemical analysis are now available so that the total iron (usually expressed in FeO) can be reported in about 15 min., while the silica-content determination takes about 25 min. The carbon content of the metal can be obtained rapidly on cast test specimens by the carbon-analyzing meters. There are also a number of indicators that serve to give a good approximation of slag composition.[13 and 14]

Any discussion of slag control must take into consideration the fundamental working of the basic open-hearth process, which places definite limitations on the conditions that can be arrived at by the use of slag control. The charge contains elements such as carbon, silicon, phosphorus, and sometimes sulphur, which must be removed to produce steel. These elements are removed by oxidation. The chief source of the oxygen required for oxidation of these elements is the flame, although some oxygen is introduced in the form of rust on the charge and by additions of iron ore during the course of the heat. These flame gases are strongly oxidizing toward the slag and steel. A schematic representation of oxidation cycles in the open hearth is shown in Fig. 1.

From a consideration of this mechanism it is evident that the rate of oxygen transfer to the bath is dependent upon the concentration in the slag of the oxygen carriers. Since slag control is aimed at governing the rate of oxidation of the bath, it is therefore concerned primarily with regulating the content of iron oxides and the consistency or viscosity of the slag.

The total iron oxide content of the slag does not by itself give a com-

plete picture of the oxidizing power of a slag. As the basicity of the slag increases during melting, the total iron oxide in the slag increases, as is shown by Herty[3] in Fig. 2, wherein silica is plotted against the total FeO content. This is a general relationship between silica and total iron oxide in the slag during melting. One steel foundry has constructed
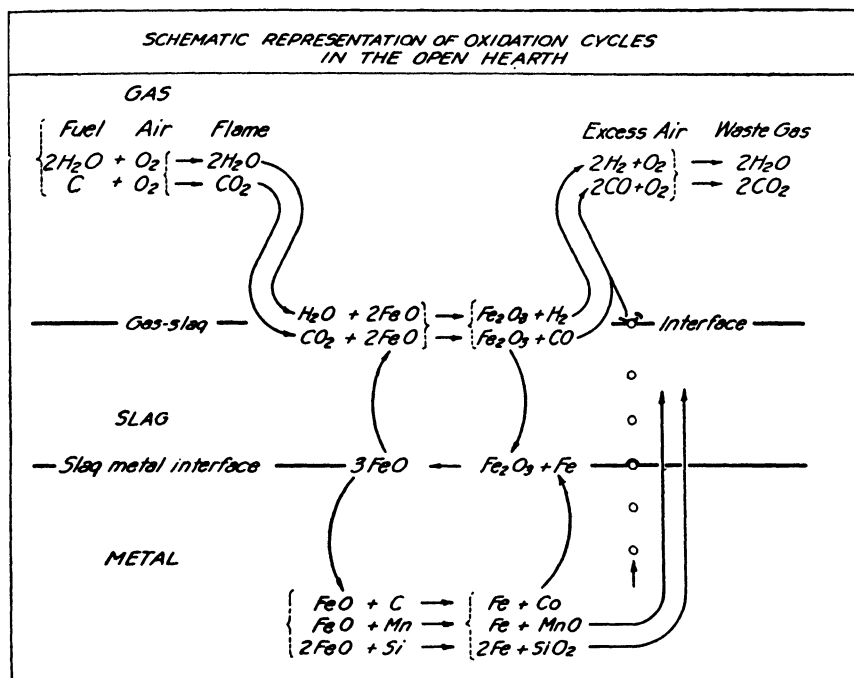

FIG. 1.—Oxidation cycle in the open hearth. (*Washburn and Philbrook.*[4])

a table based on the relationship of silica to ferrous oxide content of the slag as is shown in Table VI.

Since the carbon content of the bath may affect the relationship, more specific conditions are presented in Fig. 3 by Fetters and Chipman,

TABLE VI.—RELATION OF SILICA AND FERROUS OXIDE IN A BASIC OPEN-HEARTH SLAG

Ferrous Oxide, Per Cent	Silica, Per Cent
12–13	15–18
13–15	13–14
15–16	14–15
16–17	13–13.5
17–19	11–13

which illustrates lines of regression for total iron oxide of the slag as a function of the basicity ratio R. A regression line is calculated from an
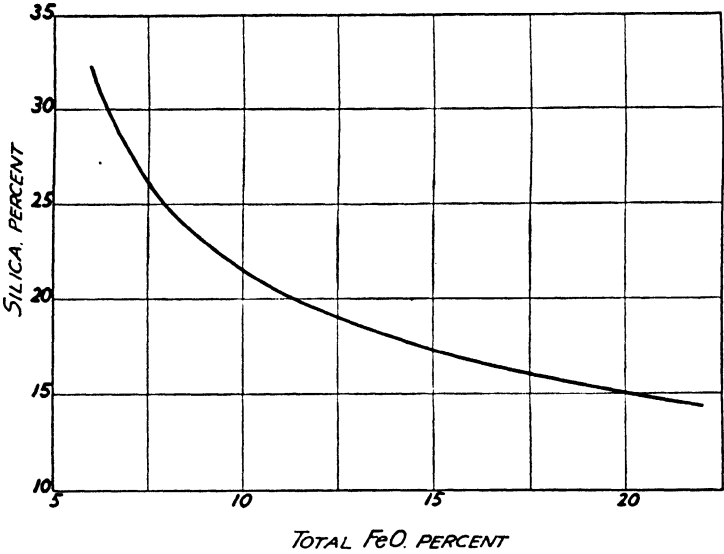
FIG. 2.—General relationship between silica and total iron oxide in slag during melting. (Herty.[3])
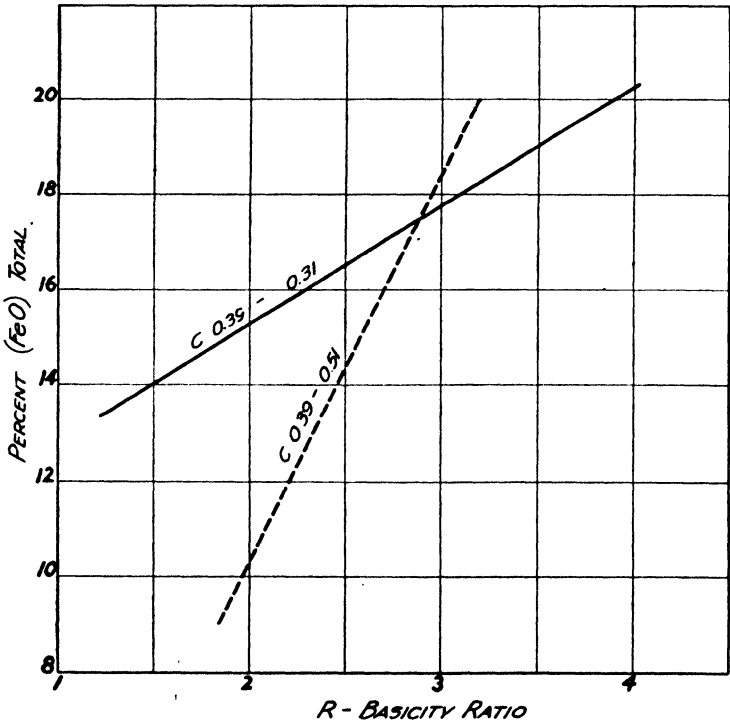
FIG. 3.—Lines of regression for total iron oxide of the slag as a function of the basicity ratio. (Fetters and Chipman.[8])

appreciable number of tests, which in this case were for the two groups of carbon contents as shown.

Much has been said about rates of oxidation of the slag and rates of transfer of FeO to the bath. The rate of oxidation of the slag by the flame is primarily dependent directly upon the partial pressures of oxidizing gases, and, secondly, it is directly proportonal to the concentration of FeO in the slag at the slag-gas interface and inversely proportional to the concentration of Fe_2O_3 at the lower layers.
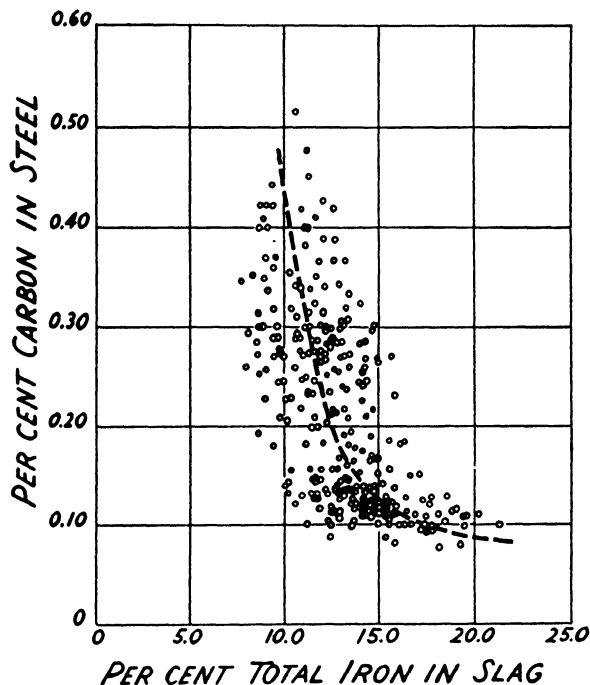
Fig. 4.—Relation between carbon in steel and total iron content of slag. Data from 300 heats of quality killed steel just before blocking. (*Washburn and Philbrook.*[4])

The concentration of active iron oxides in the slag is determined by the carbon content of the bath where the slag is sufficiently fluid. This is best shown in Fig. 4. The fluidity of the slag and the degree of agitation likewise control the rapidity of replacement of Fe_2O_3 by more FeO and thereby govern the ratio of FeO to Fe_2O_3. It may be seen, therefore, that about the only control the furnace operator has over the iron oxide factor is through the regulation of slag viscosity.

The rate of oxidation of the bath is directly proportional to the oxidizing power of the slag and inversely proportional to the concentration of FeO in the steel. Rapid solution of FeO in the steel gives rise to a more violent carbon boil. The viscosity of the slag, therefore, exerts the major

influence on the rate of oxidation of the bath by the slag. As the carbon drops, a higher concentration of FeO is necessary to bring about the subsequent drop. This is due to the fact that equilibrium conditions show that a higher percentage of FeO is present in the metal with lower carbon contents. This condition may be readily observed from Fig. 5 as prepared by Fetters and Chipman.[8]

In the same way in which the basicity of the slag and therefore the viscosity of the slag control the FeO content in the slag, it is possible to control the phosphorus and sometimes the sulphur content.
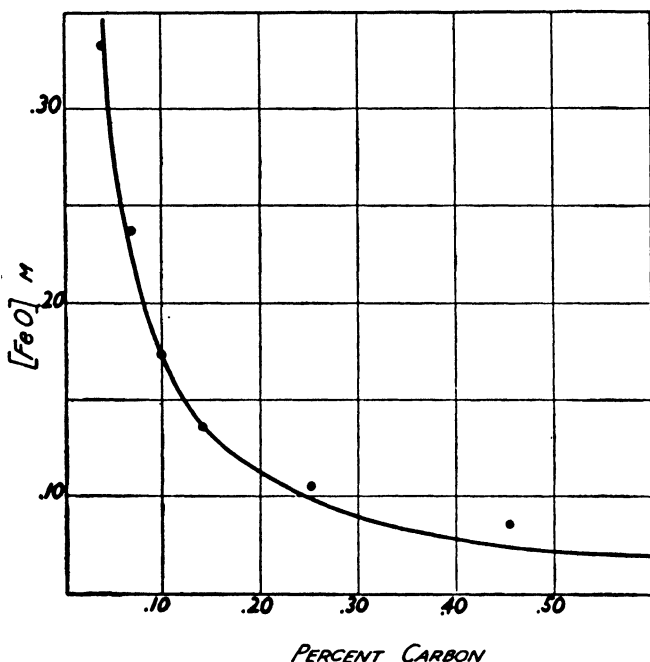
FIG. 5.—Average relation between [FeO]m and carbon. (*Fetters and Chipman.*[8])

In order that a definite plan is followed, Herty[3] has proposed a chart, Fig. 6, as an aid in making the proper corrections to slags. This figure shows the iron oxide content in the slag plotted against a slag property that indicates acidity. On the chart are shown four zones:

1. If the slag is high in iron oxide and quite basic, the addition of an acid constituent may be indicated, depending on the grade of steel being made. This acid addition will avoid the use of excessive amounts of fluorspar and will assist in keeping the iron oxide content of the slag within the desired range.

2. In another range of slag acidity and FeO, the heat will finish with the desired slag with no additions. This zone naturally will shift on the chart, depending on the grade of steel being made.

3. If the slag is exceedingly acid and reasonably low in iron oxide, a basic addition is indicated, which will give the desired lime-silica ratio at the end of the heat and assist in raising the iron oxide content of the slag in refining.

4. If the slag is extremely low in iron oxide, it may be necessary to make an addition of oxide to the slag and usually to make this addition in conjunction with a basic addition.

By following the above plan the heat will not require corrective additions during the latter part of the refining period. However, tests of the slag should be continued to ascertain that the lime-silica ratio is what
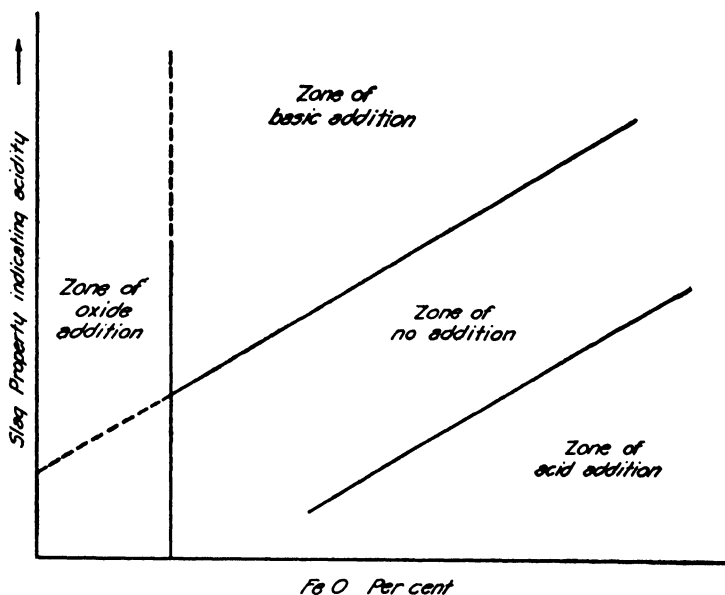
FIG. 6.—Correction chart for basic open-hearth slags. (*Herty*.[3])

is desired. At this time fluidity tests to determine this ratio are sometimes of little value because fluorspar may have been added. Slag cakes and slag color will then serve as the best guides of slag composition.

The elimination of phosphorus from the steel is one of the most important functions of the basic open-hearth practice. The controlling factors in phosphorus content of the finished steel are basicity, iron oxide content, P_2O_5 content of the slag, and temperature. The effect of basicity of the slag on the phosphorus content of the steel is shown by Washburn and Philbrook[4] in Fig. 7. A ratio of lime to silica of 2.5 during the refining period is sufficient for the removal of phosphorus for most grades of wrought steel. In fully killed steel for steel castings where drastic furnace deoxidation practices are employed, it is frequently necessary to go to a ratio of 3.0 to 3.5 to prevent phosphorus reversion

during tapping and pouring. If the FeO content is below 12 per cent, phosphorus is not eliminated readily and a reversion of phosphorus from the slag to the metal may result.

Figure 7 also shows a relationship between phosphorus and CaO/SiO_2 ratio for cast steel. The points were obtained experimentally by a basic open-hearth steel foundry.

In addition to lime and FeO, rapid removal of phosphorus from steel requires a fluid slag. Excessive basicity of the slag hinders the phosphorus removal; hence fluorspar may be necessary as an addition. The
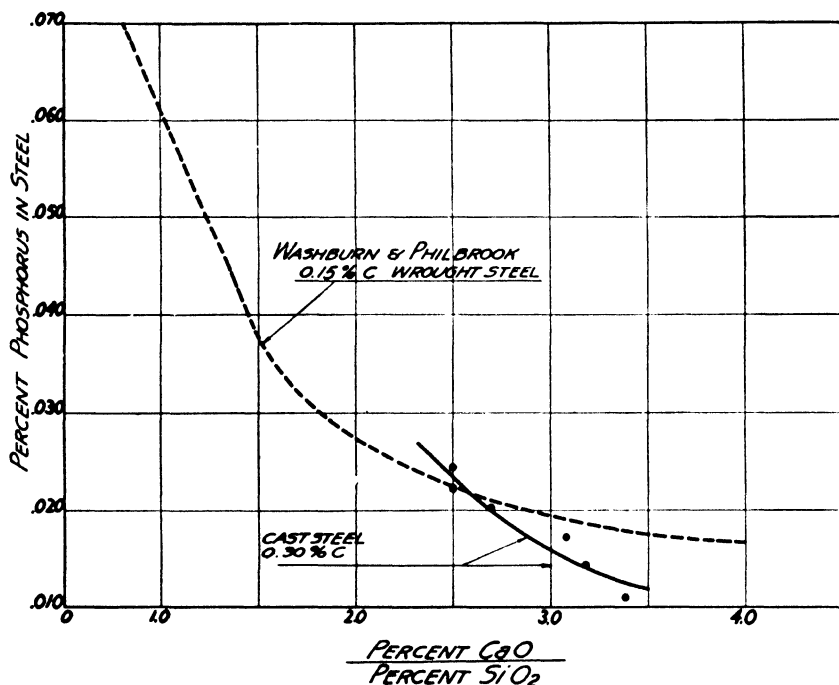
FIG. 7.—Effect of basicity of the slag on phosphorus content. (*Washburn and Philbrook.*[4])

residual-manganese content of the bath, before manganese additions are made, varies inversely as the iron oxide content of the slag. The higher the FeO content, the lower the residual-manganese content. This is shown in Fig. 8. The amount of residual manganese at any one FeO content will depend somewhat upon the MnO content of the slag. The upper curve shows high residual manganese and is encountered when the particular basic practice used calls for a high MnO content, obtained by the addition to the charge of manganese ore instead of iron ore, or of both. The lower curve is based on the usual practice of iron-ore additions. A review of a number of manufacturers making killed low-carbon steel to

phosphorus limits has reported the use of lime-silica ratio of 2.4 minimum with 2.5 to 3.0 normal operating ratios.

The amount of sulphur eliminated will depend largely upon the amount of sulphur in the charge. If the sulphur content is high, a highly
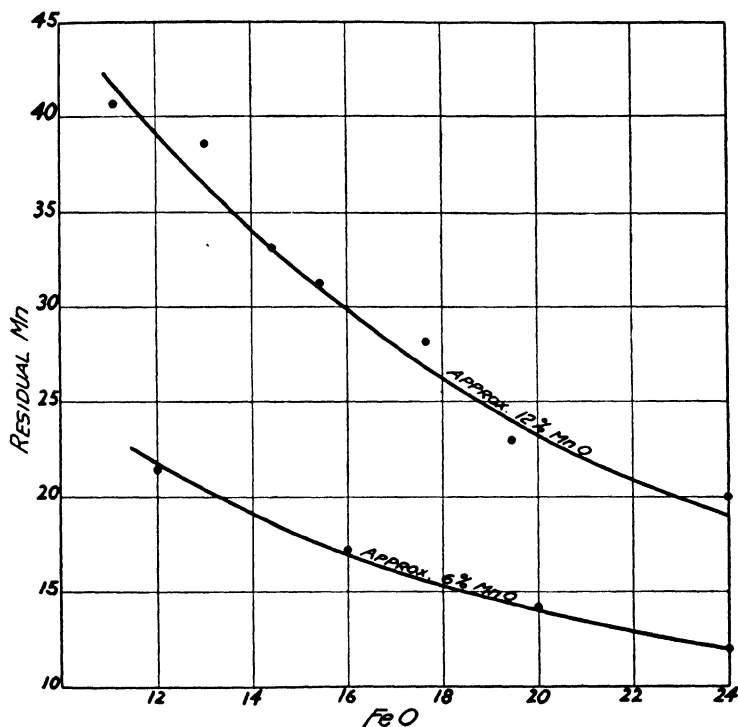

Fig. 8.—Relation between residual manganese and FeO content in basic open-hearth practice.

basic slag of CaO/SiO_2 ratio greater than 3.0 must be employed for as long a time as possible. Sulphur elimination also depends on a low FeO content. A high MnO content in the slag will make the FeO content less effective, and hence a low value of FeO/MnO is desirable.

TABLE VII.—SLAG COLOR AND COMPOSITION

Color	FeO	CaO/SiO_2 ratio
Black	Medium	Low
Gray	Low	Low
Light brown	Medium	Medium
Dark brown	Medium	High
Chocolate brown	High	High

Colors on Water Cooling.—Slag control has been rather extensively performed in basic open-hearth furnace operations. In fact, nearly all the information published on slag control is in relation to basic open-hearth slag.
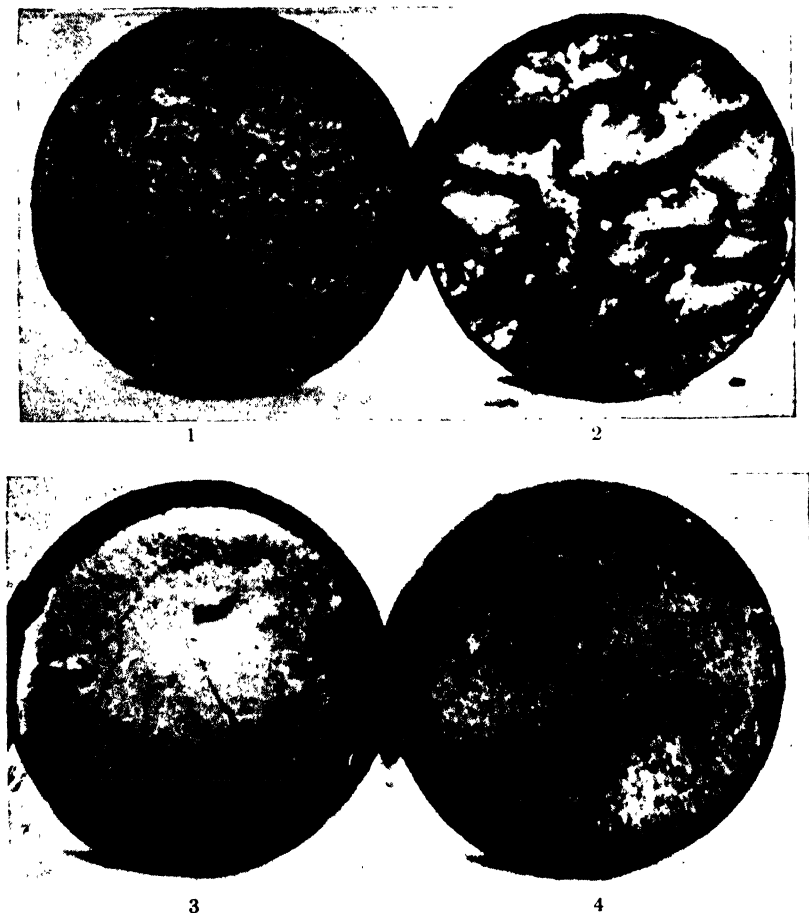

FIG. 9.—Slag cakes. (*Washburn and Philbrook.*[4])

		Lime-silica ratio
1. Creased slags	....	1.2 to 1.6
2. Furrowed Slags	....	1.4 to 2.1
3. Intermediate Slags	....	2.1 to 2.4
4. Basic slags	....	Over 2.4

The color of basic open-hearth slags when cooled in water is a guide to the approximate composition of the slag. Slag color and composition may be grouped roughly as shown in Table VII.

Slag Cake.—A more quantitative method of measuring the basicity of a slag is by visual inspection of slag cakes. Slag is poured into molds $4\frac{1}{2}$ in. in diameter and $\frac{7}{8}$ in. deep, and the surface of the resulting cake

is examined after the cake has air-cooled. The following precautions should be followed in sampling:

1. Spoons for sampling should be free from any previous slag.

2. Samples should be taken from the same location of the furnace about 3 ft. from the banks and from as near the center of the slag layer as possible in order to secure a representative sample.

3. The sample should be poured as quickly as possible into a clean test mold. Unnecessary chilling of the sample will alter the appearance of the slag cake.

4. The slag should not be allowed to run over the top of the test mold but should be filled to within about ⅛ in. of the top.

A visual examination should be made of the top and bottom surfaces of the cake and its fracture. From the top and bottom characteristics of the slag cake, it is possible to classify it into one of the following groups:

1. Creased slags
2. Furrowed slags } Acid group

3. Intermediate slags
4. Basic slags } Basic group

In Fig. 9 there are shown these four types, together with the corresponding lime-silica ratio.

After classification of the particular slag cake has been decided, the lime-silica ratio and the total iron (FeO and Fe_2O_3 reported as Fe) content of the slag can be estimated, as has been done by Mauer.[7] The four types of slag described by Mauer[7] are as follows:

Creased Slags.—The appearance of the top surface of the cake is rough or creased throughout. The bottom is shiny, with a flat or convex conformation.

Furrowed Slags.—The top surface of the cake is covered with deep furrows. The bottom is shiny, but not to such a high degree as the creased slags.

Intermediate Slags.—The top surface is black and shiny. The bottom is flat with a dull to slightly shiny luster.

Basic Slags.—The center of the cake is shiny black, while the edge is silvery or sometimes termed "gray." The bottom is flat or concave, with a dull luster. All of these slags are shown in Fig. 10.

One open-hearth foundry made a study of the color of the slag cake and worked up a chart (Fig. 11) that shows the variations from a shiny black to a dull black, which they have designated as gray. As the total FeO content of the slag increases, the surface takes on a dull-black or gray shade. This change in color does not necessarily take place in the center portion of the cake but may appear entirely to one side. A further distinction is made when the bottom of the cake appears dull, and then

the CaO/SiO$_2$ ratio is greater than 2.1, while if it is shiny, the value is under 2.1.

Melters and observers can be trained in slag-cake reading. Surface markings resulting from high MnO and high P$_2$O$_5$ give useful information on the working of the slag with respect to elimination of manganese and
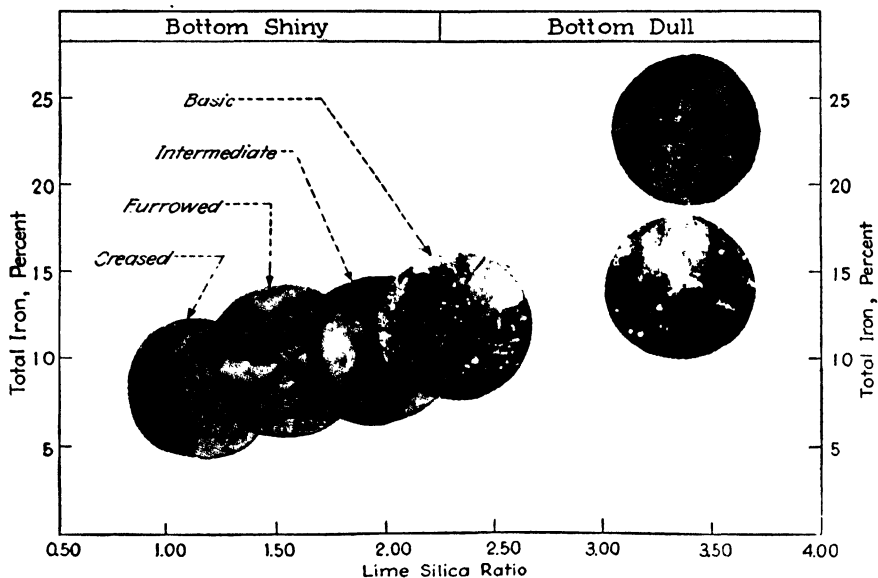

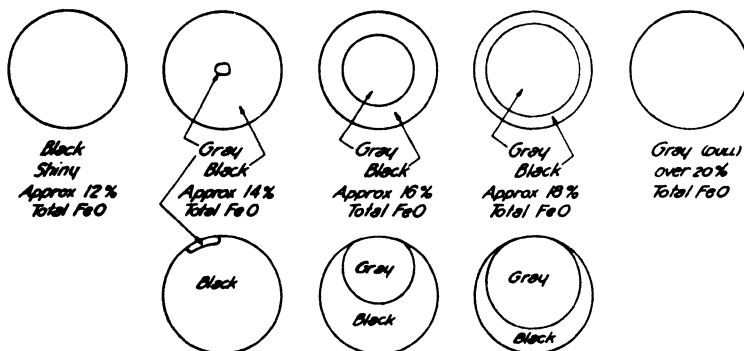Fig. 10.—Appearance of slag cakes. Total iron vs. lime-silica ratio (*Mauer*.[7])

Fig. 11.—Basic open-hearth basic-type slag cakes.

phosphorus. Changes in raw materials affect the appearance of the cakes, and, therefore, a given interpretation in one shop may not be applicable to some other shop. Each shop should set up its own standards.

Slag Viscosity.—Another useful characteristic in determining slag position is slag viscosity. Viscosity is usually determined by the flow

of slag in a tube such as the Herty viscosimeter. By the use of the viscosimeter, an accurate control of slag fluidity can be obtained. The maintaining of consistent slag fluidities for certain types of steel and for certain methods of deoxidation results in more uniform furnace practice. As a result, it is possible to hold to a more consistent slag quality, which consists of fluidity and chemical analyses, and, accordingly, it is possible to tap more heats within specification limits and to make heats more uniform in nearly all characteristics.

Figure 12 shows a sketch of the viscosimeter, which is a steel forging or casting with a funnel at one end, connected to a 0.25-in.-diameter tube, 10 in. long. It is made in two pieces, to allow ready removal of the slag sample. By filling the funnel with molten slag and by measuring in inches the distance that the molten slag runs out into the 0.25-in. tube before solidifying, an indication of the slag fluidity is obtained. Tem-
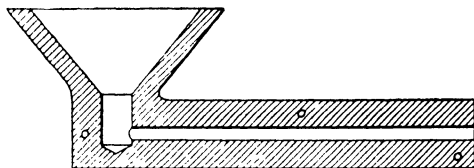

Fig. 12.—Cross section of the viscosimeter. (*Regan.*[6])

perature of the viscosimeter itself and temperature of the molten slag are, no doubt, factors, but in practice the results do not seem to be materially affected when the viscosimeter becomes warm, *i.e.*, up to the temperature that permits handling with gloves. Superheating of the slag is relative to its viscosity and should be taken into account whenever possible.

The spoons used should have a volume in excess of that required to fill the funnel cup of the viscosimeter. Spoons should be clean and free from any previous slag. The spoon should be given a single coat of slag and then dipped so that a representative sample of the slag cross section is obtained. The spoon with sample should then be withdrawn quickly and immediately dumped or inverted into the viscosimeter funnel. A steady rest may be desired for rapid location of the spoon over the funnel.

Slag samples should be taken from the same location of the furnace, care being used to obtain a representative slag sample, and cold slag or slag near the banks being avoided. A calibrated rod in inches is used for measuring viscosity, the length of the run in the viscosimeter, and may be used immediately after pouring the slag.

It has been suggested that tests be taken during the melting period and at melt-down. Additional tests should be taken during the working or refining period, before the manganese and silicon additions, and before

tapping. The length of run should be correlated against the lime-silica ratio as determined by chemical analyses.

The general relationship between slag fluidity as measured by the viscosimeter and the lime-silica ratio of basic open-hearth slag is shown by Herty[3] in Fig. 13. Changes in raw materials in the charge may shift the position of this curve slightly.

The viscosimeter and the slag cake are used throughout the melting period and the lime boil. This enables the proper corrections to be made to the slag.
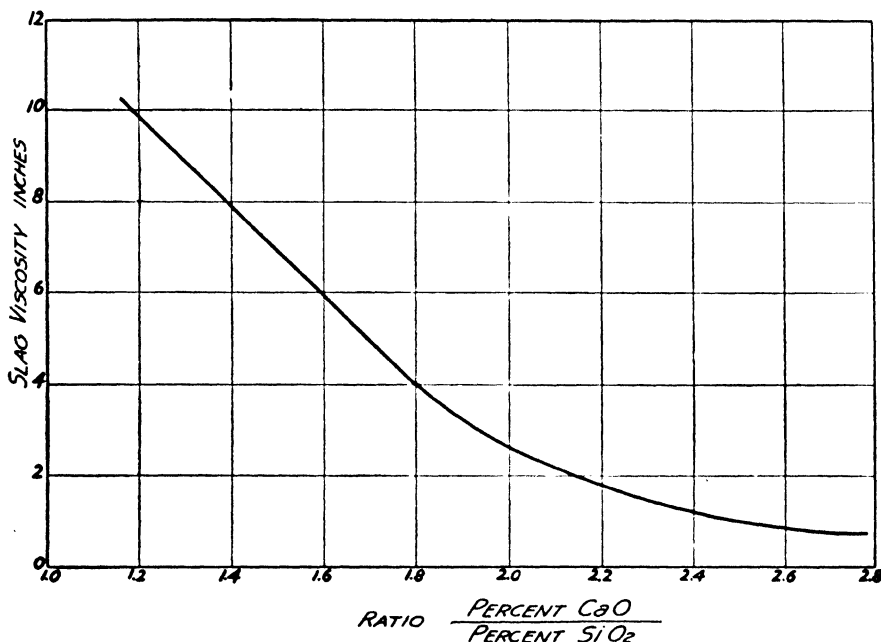

Fig. 13.—General relationship between slag viscosity and basicity during melting. (*Herty.*[3])

Tests on viscosity of slags as made by foundries operating basic open-hearth furnaces are indeed few. Most values observed were too scattered to show definite trends either as a function of the lime-silica ratio or of the total FeO content. Viscosity tests taken after the addition of fluorspar to the slag do not give values in accord with the normal viscosity expected from a basicity standpoint. Therefore, viscosity tests after the addition of fluorspar are of little value.

Steel Foundry Practice.—The basic open-hearth practice in steel foundries calls for a fairly short refining stage of 45 min. to 1 hr. for a 25-ton charge. (For 50-ton charges the refining time is approximately 2 hr.) Because the period is short, especially in the smaller size furnaces, it is feasible to alter furnace conditions by additions of sand or lime to the

bath, to effect changes in the lime-silica ratio. A more uniform control can be brought about by closely regulating the slag-forming materials in the charge. In so doing, it is necessary that special attention be given to the amount of sand impregnated on the return scrap.

By the use of a standard charge and standard slag-forming additions, it is possible to obtain in heat after heat a rather uniform degree of basicity with the desired FeO content. This is the practice followed by nearly all foundries operating on the basic open-hearth practice. Furthermore, it has been noted that early additions of fluorspar to "shape" the slag or bring it to the proper consistency result in a refining period of maximum efficiency with all the available lime in working condition. The use of fluorspar early in the refining period makes unreliable the use of the viscosity test to determine the basicity of the slag. This accounts for the fact that the viscosimeter is used so little by basic open-hearth operators in the steel casting industry. The effect of fluorspar additions can be seen in slag-cake and viscosity tests in as short a period as 10 min.

A study of a number of basic open-hearth slags in the steel foundry shows that slag samples taken just prior to the blocking of the heat will vary as follows:

	Per Cent
CaO	40–50
SiO$_2$	13–18
MnO	7–15
FeO	12–16

The above values are in some cases fairly wide in range, owing to various modifications that exist in the practice as employed at the different foundries.

BASIC-ELECTRIC PRACTICE

There were, in 1945, 57 steel foundries producing basic-electric steel for castings, in about 85 furnaces. The furnaces used vary in rated capacities from $\frac{1}{4}$ ton to 25 tons. A production of about 150,000 net tons of castings was produced in 1944 by this method. This included carbon, low-alloy, and heat and corrosion grades, and 12 to 14 percent manganese steel.

Well-made basic-electric steel has been recognized as quality steel. The process is divided into two periods: (1) the oxidizing and (2) the reducing periods. Both periods have important functions, but it is the reactions that take place during the reducing period that make the process distinctive.

In the basic-electric process the phosphorus and sulphur content can be reduced to a low figure, which, of course, is very desirable when com-

plying with specifications requiring minimum quantities of these elements. A steel with a low sulphur content is apparently not affected by low ductility upon the addition of aluminum. Numerous tests have indicated that high mechanical properties may be obtained with basic-electric steel when such critical amounts of $\frac{1}{2}$ to 1 lb. of aluminum per ton of steel are added. This apparently results from the fact that basic-electric steel usually contains less than 0.02 per cent sulphur.

The process offers considerable flexibility in the production of a large variety of steels. The method is adaptable to the making of high-alloy types as well as the regular grades. It is the preferred method for making the 12 to 14 per cent manganese steel. Large tonnages of stainless steels and corrosion-resistant steels are made by this method because of the fact that easily oxidized elements, such as manganese and chromium, can subsequently be reduced during the refining period. Care in the selection of scrap, with reference to its phosphorus and sulphur content, is not so vital a factor as in some other processes, and consequently a lower priced scrap may be used.

The basic-electric process is not used in the production of steel castings today to the degree that the acid-electric process is used. This is undoubtedly owing to the availability of the high-grade scrap of low phosphorus and sulphur content required in the acid process. Also, the relatively higher costs of basic refractories and the longer time that it takes to produce a heat of steel make the acid method attractive. It should be pointed out, however, that the use of phosphorus additions to produce high-tensile steels (rolled steels) and the high-sulphur machining stock made the scrap problem during the period of the Second World War more complex, and it is quite possible that for these reasons the basic process is finding greater application in the steel foundry.

Furnace Linings.—The furnaces in which the basic-electric process is used throughout the industry vary in rated capacities from $\frac{1}{2}$ to $7\frac{1}{2}$ tons in the average size.

The Roof.—The roof construction is usually dome-shaped, like that of a beehive brick kiln, and is of approximately uniform thickness of about 12 in. throughout. Most arc-furnace roofs are made with silica brick, but for small furnaces better results are sometimes obtained with high-alumina fire clay or sillimanite brick. Roofs have been constructed from magnesite and chrome magnesite, but the results obtained do not appear to warrant the use of these materials as yet except for small furnaces. The silica-brick roof is constructed with an expansion allowance of $\frac{1}{8}$ to $\frac{3}{16}$ in. per ft. Combustible expansion strips of wood or paper are used to provide the necessary clearance.

The life of the basic arc-furnace roof is short and variable. In some cases a life of 40 to 50 heats is considered reasonable, while under other

conditions 100 to 200 heats are expected.   Factors that contribute to the variations in roof life are intermittent operations, length of refining period, rapid heating and cooling, attack of fines from added basic materials, and types of steel produced.

Side Walls.—The side walls may be constructed of brick or may be rammed as a monolithic structure.   The former procedure is the one generally used, but where graded dolomite is available, this material has been used.   In the latter case the side walls are built with an outward slope to give added stability, and the material is rammed between a steel or wooden form and a layer of fire-clay bricks, which are placed against the furnace shell.

If the walls are brick, they are vertically faced and the bottom material is brought up high as a protection against the slag.   The 6-ton furnaces have side walls that are about 15 in. in thickness.   Expansion allowances are made by putting a small amount of insulation material between the brick and the shell.   The brick are set in a good quality of cement.

The standard lining is constructed from magnesite brick or metal-encased magnesite brick.   More recently magnesite brick have been replaced by chrome-magnesite brick, which offer the advantage of lower conductivity and higher thermal shock resistance.

The wall life varies from 75 to 100 heats, or it frequently is based on the life of the roof.   Damage results mostly from spalling and from attack by iron oxide present in the melting slags.

In some cases, because of the cost of magnesite brick, these brick are used for walls to a point only slightly above the slag line.   On top of this are placed one or two courses of chrome brick, and above these are laid silica brick.   In some cases the layer of chrome brick is not used.   The life of such a wall is not so long as the completely basic wall, and it is doubtful if there is much saving, if any, in the cost of a wall lining.

An expansion allowance of approximately $\frac{1}{4}$ in. per ft. is normally provided for magnesite brick back from the metal area.   At the metal face an expansion allowance of $\frac{1}{16}$ to $\frac{1}{8}$ in. is allowed between the bricks.

Bottom Lining.—The hearths are laid in several ways.   In some cases a fused hearth is built up, while in other cases a rammed hearth is used. The rammed hearths are becoming increasingly more popular.   Considerably less time is taken in the preparation of a rammed hearth, and the economy effected and the efficiency and serviceability obtained are such that the more progressive organizations have switched over to the rammed-in hearths.

The bottoms or hearths of the larger furnaces have in the past usually been constructed from graded, calcined magnesite or dolomite that is fused into place.   Such a construction is generally found when consider-

able amounts of iron ore are used in the melting practice.   The subhearth is insulated with a flat course of firebrick against the shell, on top of which are laid several courses of magnesite brick extending up the side walls to just above the slag line.   The side walls and roof are lined with silica brick.

The accepted practice is to place three pieces of used or broken electrode on the bottom so that they will be under each furnace electrode and in contact with each other.   The furnace electrodes are then lowered and the current turned on.   A low tap (low voltage) should be used at first, so that the furnace will be heated gradually and allowed to dry out. When the brickwork is thoroughly dry, the temperature may be brought up to the sintering heat and the first layer of magnesite may be placed. When this has sintered tightly to the brickwork, the next layer is shoveled in and the process continued until the bottom is complete.   .

Usually a small amount of ground slag (basic open-hearth or first slag from the basic-electric) is added to the magnesite to help the magnesite sinter in place.   Small additions of ground calcined dolomite are sometimes substituted for the slag, since the dolomite has a sintering point that is lower than that of magnesite.   Typical analyses of magnesite and dolomite are shown in Table VIII.   The pieces of electrode used for resistors are pried up as the bottom is built up.

TABLE VIII.—TYPICAL ANALYSIS OF MAGNESITE AND DOLOMITE

	Per cent	
	Magnesite	Dolomite
SiO_2	4 –8	12.0
Al_2O_3	2	4.0
Fe_2O_3	4 –8	5.0
CaO	2.5–7	47.0
MgO	80	32.0

In all the small furnaces and in most of the larger ones the bottom is rammed in place.   One method requires the use of periclase (90 per cent MgO with less than 5 per cent SiO_2), which is mixed with a high-temperature bonding agent such as Austrian magnesite.   To this mixture is added about 3 per cent sodium silicate to make the mix sufficiently moist to allow the material to be tamped.   The periclase is well graded so that about 75 per cent passes the 6-mesh sieve and remains on the 20-mesh. The mix contains about 70 per cent of this material, and the other 30 per cent is made up of equal amounts of finely ground periclase and the high-temperature bonding agent.   The moistened material is rammed to the shape of the desired hearth.

Another material that is being used extensively is Ramix, which is made from stable magnesia clinkers. The clinker contains from 65 to 82 per cent magnesia, depending on the grade required. The clinker is carefully sized and a chemical bond is combined with it. These bonds give a strong "air set" to the refractory when it is mixed with water to the consistency of molding sand. Rammed bottoms of 4 to 6 in. are normally prescribed for the furnaces of 6 tons and under, whereas a 6- to 8-in. lining is sufficient for larger furnaces.

Rammed bottoms may be brought into operation within 24 hr. A wood fire placed in the closed furnace will dry out the brickwork in about 6 hr. This is often followed by a coke fire for approximately 6 hr. to heat the furnace, after which the current is turned on and the electrodes are allowed to arc on the coke. The temperature of the furnace is brought up and held for several hours at a temperature just below that which would cause the silica brick in the side walls and roof to drip, or about 3000°F. This treatment will sinter the top surface of the hearth. In some cases, after the furnace is dried, the heat is charged and melted prior to sintering the surface of the hearth. A still further modification of rammed bottoms consists of magnesite, or magnesite and dolomite, to which is added hot tar for tamping purposes.

Raw Materials.—The slag-making materials for the basic process consist of lime, sand, fluorspar, and a carbonaceous material such as ground carbon, graphite, or coke. Analyses of these materials are given in Table IX.

The lime should be of good metallurgical grade and contain not less than 95 per cent CaO. Excellent grades of lime may be obtained if calcined lime of the analysis as shown in Table IX is specified. The material should be purchased in rice-to-nut size with the dust screened out. If possible, the lime should be obtained in airtight containers since the lime breaks down into fines when standing in the air.

TABLE IX.—COMPOSITION OF SLAG-FORMING MATERIALS

	Per cent				
	CaO	SiO$_2$	MgO	CaF$_2$	Fe$_2$O$_3$ + Al$_2$O$_3$
Lime................................	95.0	1.0	1.5	....	1.5
Sand................................	...	98.0	...	....	1.5
Fluorspar..........................	5.0	5.0	...	85.0	5.0

Limestone is seldom used in place of metallurgical lime, partly because it has to be calcined in the furnace, which delays operations, and partly because it takes longer to form a slag of the proper consistency in the

refining slag. Also, the oxygen content is higher in those steels where limestone is used in the refining slag, thus causing an increase in the consumption of deoxidizing agents. However, limestone is sometimes used during the melting period and in the forming of the oxidizing slag.

The primary function of fluorspar is to lower the viscosity of the basic slag. It also aids indirectly in the removal of sulphur from steel by decreasing the slag viscosity. Fluorspar makes it possible for a slag to have a high lime content without being excessively viscous—a fluid slag being more reactive than a viscous one—and it does this without changing the basicity of the slag. Silica sand is also used occasionally as a thinning agent. Addition of sand, however, lowers the basicity of the slag.

The Charge.—The charge consists of foundry scrap (gates and risers) and purchased scrap. From 20 to 50 per cent will consist of foundry scrap. The purchased scrap will be of numerous varieties. A charge containing over 0.10 per cent phosphorus and 0.08 per cent sulphur is seldom used, and generally it will run considerably below this figure. If the charge contains scrap that is low in carbon content, an addition of 2 or 3 per cent of pig iron is made so that sufficient carbon is present to produce a boil after the charge is melted.

The furnace is charged first with the heavy scrap, which is placed directly under the electrodes. Light scrap is charged around the electrodes and in the center of the furnace. The light bulky scrap, such as turnings, is charged around the sides. Limestone or lime is sometimes charged with the scrap, but care is taken to keep these nonconducting materials from interfering with the formation of the arc.

The placing of the scrap in the furnace and the proper proportion of light and heavy scrap are important from the standpoint of bottom difficulties and rapid melting. The heat should melt from the bottom up. If the arc is concentrated mostly on the heavy scrap, a pool of metal will be formed quickly and a constant arc can be maintained. Such a practice cuts down power surges and decreases melting time.

The foundry returns (gates, risers, rejected castings) should be free of excess sand, since adhering sand will lower the basicity of the slag and thus affect the efficiency of the phosphorus elimination. The modern furnace is equipped with a removable roof, which saves considerable time in charging. However, hand charging has the advantage that scrap can be placed in the furnace in the manner selected. If charging buckets are used, care must be exercised in placing the scrap in the bucket so that it will fall correctly into the desired place.

It is considered essential to have the metal charge and all alloys, as well as all fluxes and slag-forming materials, absolutely free from moisture to prevent the formation and subsequent decomposition of water vapor in the furnace. The hydrogen so formed may be absorbed by the bath

with the possibility of producing defective steel at the time of casting. Some operators go so far as to preheat all their materials before putting them into the furnace.

Melting—the Oxidizing Period.—The manufacture of basic-electric steel castings is carried on by the cold-scrap process. In the cold-scrap process the melting operation is carried out as rapidly as possible. As soon as a pool of metal is well formed, a small amount of lime or limestone is added to protect the metal from overoxidizing. From time to time during the melting, other additions of lime or limestone are made so that when the charge is completely melted, a slag in the proportion of about 30 to 40 lb. to the ton of metal is present. This will require an addition of CaO equivalent to about 1.5 to 2 per cent of the charge.

If it is necessary to use iron ore to create a vigorous boil, to assist in the carbon, hydrogen, and nitrogen elimination, the ore is usually added just before melting is completed. In some cases it is added in two separate batches, *viz.*, just before and again shortly after melting is completed. In one plant manganese ore is used instead of iron ore.

Under certain circumstances it may be necessary to add a little fluorspar to the slag when the SiO_2 content is very low, or when no ore is added to the bath, in order to lower the viscosity of the lime slag. This, however, is not necessary if a good oxidizing slag containing about 15 per cent FeO is maintained.

As soon as the bath is melted, the power input is reduced in order that low temperature will be maintained as an aid in the phosophorus elimination, and a metal test for the chemical laboratory is taken. The slag will appear very black and will be quite brittle.

In Table X are presented the analyses that have been obtained from a study of basic oxidizing slags produced in commercial practice. The table shows the close control of oxidation that may be obtained. Since

TABLE X.—ANALYSIS OF BASIC OXIDIZING SLAGS

	A	B	C	D	E	F	G	H
CaO	38.56	50.28	43.85	51.20	45.10	36.71	17.26	30.47
CaF₂	.....	12.62	.....	.....	.....	.....	.....	0.55
CaS	0.41	0.32	0.40	0.32	0.34	0.37	0.09	0.22
CaC₂	0.13	0.18	0.10	0.09	0.13	0.24	0.10	0.06
FeO	24.62	15.50	18.02	14.85	12.25	24.12	13.80	24.55
MnO	11.68	2.90	8.93	5.52	12.34	8.49	13.30	10.32
Al₂O₃	1.96	1.07	2.16	2.94	2.02	3.45	5.24	2.00
MgO	8.29	7.56	10.00	7.76	10.70	9.94	19.83	9.39
Cr₂O₃	0.62	0.35	0.22	0.57	.....	4.26	.....	1.22
P₂O₅	1.52	1.57	2.44	0.47	1.96	0.82	0.06	0.58
SiO₂	12.69	6.86	13.82	15.39	15.17	12.38	29.62	21.04

during the melting period there are no oxidizing gases present as in the open-hearth process, the oxidation may be controlled by the amount of oxygen (in the form of a solid oxide) that is admitted to the furnace. The solid oxide may be present in the form of iron ore, mill scale, or rusty scrap.

It will be noted from the table that the FeO content of the oxidizing slags varies from 12 to 25 per cent. Thus slags A, F, and H were melted with iron ore added to the bath. Slags of this type indicate that complete oxidation was obtained during the melting period and that carbon, manganese, chromium, and silicon were almost completely oxidized from the molten metal. If a charge is made up of scrap that will average between 0.20 and 0.25 per cent carbon, it is possible to obtain, by melting with complete oxidation, a carbon content of 0.03 to 0.07 per cent, manganese from 0.05 to 0.10 per cent, and silicon from 0.01 to 0.05 per cent. If the carbon content of the scrap averages 0.40 to 0.60 per cent, the carbon reduction will not be carried so low but will be in the neighborhood of perhaps 0.10 to 0.16 per cent.

Oxidizing slags that contain FeO in the neighborhood of 13 to 18 per cent are occasionally recorded in steel casting manufacturing. Under these conditions iron ore is not added to the bath, mill scale is kept from the charge, and the scrap is relatively free from excessive rust. This method of melting is spoken of as "melting with partial oxidation." The carbon content of the scrap before melting averages from 0.20 to 0.40 per cent, whereas at the end of the oxidizing period it will be found to fall between 0.10 to 0.20 per cent. The manganese content usually falls between 0.10 to 0.20 per cent, and the silicon is in the neighborhood of 0.05 per cent.

A boiling action will be noticed taking place across the bath during the melting period and after melting is complete. The boiling action is generally light, but upon addition of iron ore it becomes quite vigorous. The amount of ore used is varied, depending upon the degree of oxidation of the scrap, the size of the furnace, and type of steel to be produced.

The length of time that the oxidizing slag is maintained on the bath will depend on the phosphorus content and the carbon content of the bath. No ore would probably be needed if the charge contained an average carbon content of 0.20 per cent and a phosphorus content of 0.05 per cent, and if there were a normal percentage of rusty scrap in the charge to produce a slag containing approximately 15 per cent FeO; slagging-off operations could proceed shortly after melting was complete. If phosphorus were high or if carbon were high, a more oxidizing slag might be required for a longer working period. In this case, iron ore and additional lime might be added to the bath and an oxidizing slag maintained for 20 to 30 min. after melting was complete.

There is a third variation of oxidation control in which an attempt is made to melt without oxidation. This practice is not used in the normal production of carbon and low-alloy steel castings. It is, however, practiced in the manufacture of high-alloy steel castings when scrap of high alloy content is used. This is especially true when the scrap is high in chromium or manganese. Under these circumstances no ore or scale is employed and the scrap is carefully selected so that it is as free as possible from rust and scale. Lime is used in place of limestone in the slag, and the slag is not removed but becomes a base for the refining operations. In this case, of course, the small amounts of manganese, chromium, and phosphorus that are oxidized during melting are reduced to their respective elements and returned to the metal during the refining period. Only a small percentage of the carbon in the original scrap is lost.

Several other interesting points may be mentioned in connection with the slag analyses of oxidizing slags of Table X. Slag B has very basic characteristics, so much so that it was necessary to add fluorspar in order to decrease the viscosity of the slag. When fluorspar is added, in order to have the analyses total 100 per cent, it is necessary to analyze for CaF_2. This is an important point and should be borne in mind by those analyzing basic slags, especially the refining slags where fluorspar is more commonly used.

A good basic oxidizing slag will contain between 45 and 70 per cent strong bases ($CaO + MgO$), between 15 and 35 per cent weak bases ($FeO + MnO$), and between 10 and 20 per cent acids ($SiO_2 + P_2O_5$). The CaO content should be at least approximately 40 per cent. A lime-silica ratio of 2.2 to 2.8 is desirable.

Slag G is a very poor basic oxidizing slag. The SiO_2 content is about 10 per cent too high for normal practice. The high MgO content indicates that the magnesite lining is being unduly corroded because of the acid characteristics of the slag. The analyses show that the slag has little phosphorus-eliminating ability because of the low lime content.

In order to obtain an elimination of phosphorus from the bath, it is necessary that the slag contain an excess of bases and of iron oxide. Table X confirms this in that slags high in bases and iron oxide show considerable amounts of P_2O_5 present. It should be pointed out, for the sake of clarity, that slag D contained only a low percentage of phosphorus in the scrap (approximately 0.038 per cent) and, as will be shown later, was subjected only to the single-slag process. It is rather difficult to obtain an accurate determination of the amount of phosphorus that is eliminated from the bath during the melting period and the subsequent time that the oxidizing slag remains on the bath. This is because it is difficult to obtain an accurate determination of the phosphorus content in the scrap.

Also, a determination of the phosphorus content of metal after the steel has melted down is not sufficiently informative, since phosphorus elimination is proceeding during the melting-down period. A rough idea of phosphorus elimination, which may be of assistance in the planning of future heats, may be obtained by a study of the P_2O_5 content of the slag and the phosphorus content of the metal at slagging-off time. The P_2O_5 could be reconverted if the total slag weight were known. Unfortunately, under commercial practice, the weight of the oxidizing slag is not determined, and only a rough average of from 30 to 40 lb. of slag per ton of metal is maintained. Values other than this may result. depending on the erosion of bottom material, the amount of sand adhering to the scrap, the surface area of the bath, etc.

At the time the oxidizing slag is ready to be removed, the phosphorus content of the bath will be in the neighborhood of 0.01 to 0.03 per cent, provided that a good dephosphorizing slag is employed. There will also be a slight loss in sulphur content. Tests made on the charging of a uniform grade of screw stock have shown a reduction of 0.008 per cent sulphur at the time the oxidizing slag was removed.

The bath is now ready for slagging off. The double-slag process will be discussed first, after which the single-slag method and other modified procedures will be outlined.

Double-slag Procedure.—The first slag, or oxidizing slag, is removed by tilting the furnace and raking off the slag, usually through the charging door. The power is cut off and the electrodes are raised to facilitate slagging operations. Considerable care is given to the slagging operation so that all the slag may be removed from the bath. In some cases, lime is added just before slagging is completed, so that the remaining thin slag will be thickened considerably, thus allowing the melter to remove the last traces of the slag. The complete removal of the oxidizing slag is necessary since this slag is carrying phosphorus and under the reducing conditions, which are to follow, the remaining phosphorus will again revert to the metal unless these precautions are followed.

To the slag-free bath there usually is added either ferromanganese or ferrosilicon, followed by the addition of carbon. The deoxidizer is added to reduce the oxygen content of the bath. By following this procedure it is possible to assume a definite carbon pickup by the bath. If the steel is to finish at 0.25 per cent carbon, a carbon addition will be made so that the bath will contain between 0.20 and 0.22 per cent carbon. The open bath will pick up the added carbon quickly, and immediately thereafter the refining slag is added to the bath.

Another practice, and probably a better one, is to eliminate the deoxidizing additions and reduce the oxygen content of the bath by the carbon addition. Carbon is added in the form of low-phosphorus pig

or wash metal well broken up so that a general vigorous boiling action is obtained across the entire bath. A definite carbon pickup of the bath can seldom be assumed if this method is used. It is not necessary that this practice be carried out on the open bath, but it may be accomplished soon after the refining slag-forming materials are added to the bath.

In some plants the addition of ferrosilicon or ferromanganese to the open bath is considered poor practice because finely divided silicates, which are the product of deoxidation reaction, are formed. Those who practice this method claim that these reaction products will be eliminated from the metal by migration to the slag during the hour-long refining period.

The refining slag is usually mixed up beforehand and will contain a definite proportion of lime and fluorspar. About one-half the total slag material to be used is charged first. The electrodes are lowered and the current is turned on as the slag materials are added. A lime-fluorspar ratio of approximately 8 to 1 is ordinarily used. In some foundries sand is added to the mix or substituted for part of the fluorspar. Usually sand is added only with the first addition of slag-making materials, so that an easily fusible slag will be obtained as quickly as possible. As the materials fuse, further additions of lime and spar are made until the proper volume of slag is obtained. To these succeeding additions, pulverized carbon is added so that reducing conditions may proceed at once and thus prevent the oxidation of the carbon in the metal.

Calcium carbide may be one of the constituents of the refining slag-forming materials. Quantities of one-half to two-thirds of the slag material required may be used. By its use the formation of the carbide slag is speeded up. The metal-to-slag ratio is normally between 30 to 1 and 40 to 1 (2.5 to 3.4 per cent of the metal charge), depending on the size of the furnace or, stated more correctly, depending on the surface area of the bath.

As soon as the second slag is melted, the current is reduced and the refining operation proceeds. The slag color at this state is quite black. At regular intervals a small amount of pulverized carbon or coke is spread over the bath. In some foundries a small amount of pulverized ferrosilicon is added with the carbon, especially at first, to hasten the reducing action. As this reaction progresses, slag samples quenched in water, which at first may be a dark greenish-brown, turn to a lighter brown color and then disintegrate or crumble into a white or gray powder in air.

Slag samples should be taken at regular intervals during the period in which pulverized-carbon additions are made to the slag. The slag at first will foam when the carbon is added, and later, when deoxidation is nearly complete, it will have the general appearance of granular material floating on top of the bath. Just prior to the time that the slag sample

disintegrates, an odor of acetylene from the calcium carbide may be detected in the steam rising from a slagged spoon that has been placed in water.

It is possible to carry the deoxidation to such an extreme that sufficient calcium carbide is formed to break down the slag sample in air into a dark-gray powder, instead of the usual white powder. Slags of this type will contain as much as 1.50 to 3.00 per cent calcium carbide. In some foundries the white-powder slag is not allowed to form, and the light-brown slag that emits a faint odor of acetylene is used instead as the finishing slag. The reason given for using this procedure is that the steel under this type of slag has a greater fluidity than one produced under a white finishing slag. Such a slag will contain iron oxide in quantities between 1.5 and 3.0 per cent.

The period in which the refining slag is on the bath will total from 1 to $1\frac{1}{2}$ hr. Half of this time is taken up in forming the slag and the other half in deoxidizing it. Generally, $\frac{1}{2}$ hr. beyond the time the slag is made is sufficient to deoxidize the slag thoroughly. The oxides of iron and manganese in the slag are reduced to metallic manganese and iron by the reaction of carbon on the oxides and the formation of CO. The sulphur is converted to calcium sulphide and retained by the slag or escapes as SO_2. These reactions will be explained later in detail.

During the refining period the consistency of the slag should be carefully observed. The slag should not be allowed to become too viscous, since sulphur elimination does not proceed efficiently when viscous slags are used. Also, it should not be allowed to become too fluid or thin, since it then will not develop into the proper reducing slag. The latter condition is usually an indication that the SiO_2 content is becoming too high, owing either to spalling of the silica refractories or to too much sand in the slag-forming materials. It should, of course, be remembered that in order to have efficient sulphur elimination the lime-silica ratio must be kept high. The color and the appearance are good gauges of the deoxidizing value of the slag. The odor coming from the slag also indicates the presence of calcium carbide. A sample for the chemical laboratory is obtained immediately upon the appearance of the carbide slag. As soon as the carbon and manganese contents of the bath are known, preparations are made for deoxidation and tapping of the furnace.

If further additions of carbon to the metal are needed, they are as a rule added in the form of wash metal (3.5 per cent carbon with low phosphorus, sulphur, and silicon content) or a low-sulphur pig iron. The power input is increased and the metal temperature and its flowing power are adjudged by spoon set tests or fluidity tests. Upon obtaining the desired temperature, the final additions in the form of ferromanganese and ferrosilicon are made and the bath is rabbled (stirred) and tapped.

TABLE XI.—RECORD OF A DOUBLE-SLAG HEAT OF A 6-TON FURNACE

Analysis desired: C, 0.25 per cent; Mn, 0.65 per cent; SiO, 0.30 per cent; P, less than 0.03 per cent; S, less than 0.03 per cent.

Charge:	Lb.
Miscellaneous scrap	13,000
Foundry returns	5,000
Total	18,000

Time	Activity
10:00 A.M.	Power on high voltage
10:30	Lime 100 lb.
11:00	Lime 200 lb.
11:20	Lime 100 lb.
11:40	Lime 50 lb., intermediate voltage
12:05 P.M.	Melted down
12:10	First laboratory test for C and Mn
12:25	Test reported: C, 0.15 per cent, Mn, 0.13 per cent
12:30	Second laboratory test for P and S
12:32	Slag sample, black
12:35	Slag off
12:42	Finish slag off
12:43	Add: Ferromanganese 105 lb., 80 per cent Mn
	Ferrosilicon 25 lb., 50 per cent Si
12:44	Add: Pulverized carbon 20 lb.
12:45	Add: Refining slag mix:
	200 lb. lime
	25 lb. fluorspar
	20 lb. sand
12:58	Add: Refining slag mix:
	200 lb. lime
	25 lb. fluorspar
	10 lb. pulverized carbon
1:05	Second slag sample
1:10	Third metal test for laboratory, carbon
1:15	Add: pulverized carbon 4 lb.; pulverized FeSi 2 lb. to slag
1:20	Add: pulverized carbon 2 lb.; pulverized FeSi 2 lb. to slag
1:25	Carbide slag, third slag sample
1:30	Second metal test reported: P, 0.019 per cent; S, 0.047 per cent
	Third metal test reported: C, 0.22 per cent
1:30	Add: pulverized carbon 2 lb.
1:35	Add: pulverized carbon 2 lb.
1:40	Fourth slag, disintegrated, white
1:42	Fourth metal test for C and Mn
1:55	Fourth metal test reported: C, 0.23 per cent, Mn, 0.38 per cent
2:00	Stir bath; temperature test
2:05	Add: Ferromanganese 70 lb., 80 per cent Mn
	Add: Ferrosilicon 95 lb., 50 per cent Si
2:10	Stir bath: temperature test
2:12	Fifth slag test
2:15	Tapped

Ladle analysis: C, 0.24 per cent; Mn, 0.66 per cent; Si, 0.30 per cent; P, 0.020 per cent; S, 0.023 per cent.

In some cases ferrotitanium is added to the furnace just prior to tapping. Also, in other plants, a small amount of aluminum (1 to 2 lb. per ton) is added to the ladle during the tapping.

In most cases the metal is tapped into bottom-pour ladles. Top-pour ladles are seldom if ever used since it is rather difficult to hold back the slag during the pouring of molds because of its very fluid nature. Occasionally, the basic steel is hand-shanked,[1] but this procedure is an exceptional one, for basic-electric steel is considered to have rather poor shanking qualities. This means that the fluidity characteristics of basic-electric steel are considered to be inferior to those of acid-electric steel. Considerable basic-electric steel has been hand-shanked, but for this purpose the steel is practically always made by the single-slag method.

TABLE XII.—SLAG ANALYSIS FROM HEAT OF TABLE XI. DOUBLE SLAG. CARBON STEEL

	Per cent				
	Oxidizing slag	Refining slag			Tapping slag
	12:32 P.M. No. 1	1:05 P.M. No. 2	1:25 P.M. No. 3	1:40 P.M. No. 4	2:12 P.M. No. 5
CaO............	41.02	43.28	49.91	51.84	51.35
CaF_2....	......	6.98	7.30	7.96	7.05
CaS............	0.22	0.28	0.41	0.46	0.71
CaC_2............	......	0.16	0.70	0.58	0.59
FeO............	23.90	11.55	0.88	0.43	0.67
MnO.....	7.00	8.42	2.96	0.91	0.30
Al_2O_3............	1.96	1.61	1.76	1.50	2.12
MgO....	9.56	7.69	9.08	7.42	9.60
Cr_2O_3............	0.75	0.56	0.08	0.02	0.02
P_2O_5............	0.98	0.32	0.21	0.07	0.02
SiO_2............	14.56	19.72	27.68	29.30	27.18
Color............	Black	Dark brown-green	Light brown	White	White

A furnace record of a typical double-slag heat is presented in Table XI. From a study of the miscellaneous scrap it is believed that the phosphorus and sulphur contents were approximately 0.05 per cent each at the time melting started. Slag analyses were also made on this heat and are given in Table XII. Attention is called to the manner in which the

[1] Hand-shanked—a term used to describe steel when small castings are poured by a man carrying a ladle holding up to 50 lb. of steel.

FeO content drops and the CaS content increases as the refining time continues.

In Table XIII are given other typical refining slags at tapping time. These analyses bring out the fact that the SiO_2 content varies from 15 to 30 per cent and the FeO content will be present in amounts of 0.50 to 1.50 per cent, which is considerably lower than that found in the oxidizing slag.

TABLE XIII.—SINGLE-SLAG METHOD. TYPICAL FINISHING SLAGS AT TAPPING TIME

	Per cent				
	No. 1	No. 2	No. 3	No. 4	No. 5
CaO	45.23	51.39	54.87	54.26	43.42
CaF$_2$	11.05	11.52	13.53	18.50	6.69
CaS	0.53	1.82	0.41	1.35	0.23
CaC$_2$	0.50	1.05	1.03	1.22	0.28
FeO	1.05	0.85	0.84	0.48	2.50
MnO	0.73	0.15	0.42	0.35	2.88
Al$_2$O$_3$	2.72	3.05	1.20	2.06	3.15
MgO	14.53	13.02	6.56	7.77	8.50
Cr$_2$O$_3$	0.02	. . . . .	0.02		
P$_2$O$_5$	0.03	0.02	0.06	0.03	0.02
SiO$_2$	23.68	17.42	19.80	14.04	31.06
Color	Light brown	White	White	White	Dark brown

A fluid basic refining slag will lower appreciably the sulphur content of the metal. It is not uncommon to obtain a sulphur content in the neighborhood of 0.010 per cent after good refining conditions have been maintained. Table XIV lists a few analyses showing sulphur percentage present in the metal both before and after the refining period. Reduction in sulphur of 0.10 to 0.01 per cent requires the extension of the refining periods to much longer ones than those usually given in normal operation or requires a slag containing 2.00 to 3.00 per cent calcium carbide throughout the refining period. With large percentages of calcium carbide present in the slag, the sulphur will be quickly eliminated.

TABLE XIV.—PER CENT SULPHUR PRESENT IN METAL BEFORE AND AFTER REFINING PERIODS

	Per cent				
	1	2	3	4	5
Before refining	0.038	0.047	0.044	0.030	0.062
After refining	0.013	0.023	0.014	0.008	0.029

Constant attention must be given to temperature control in electric-furnace steel production, since the equilibrium between oxygen in the bath and in the slag changes materially with temperature. Hydrogen is also more likely to be absorbed at higher temperature; this is particularly serious when certain of the chromium-nickel steels are being melted, especially those containing considerable silicon. Hydrogen causes porosity in castings; consequently, every effort should be made to maintain an atmosphere of carbon monoxide in the furnace, particularly during the elimination of sulphur, which requires a high temperature for efficient elimination.

Petrographic analysis of a carbide slag indicates the presence of dicalcium silicate ($2CaO \cdot SiO_2$) and usually tricalcium silicate ($3CaO \cdot SiO_2$), a small amount of solid-solution phase consisting of magnesium oxide, manganese oxide, and ferrous oxide, undissolved magnesium oxide, and possibly some free lime.

Slag Control.—The control of basic slags has resulted almost entirely in correlating the slag composition with slag appearance. Also, the foamy appearance of the slag while in the furnace is a rough indication of its basicity. Slag samples are taken on spoons and dipped into cold water and then observed. The color indicates the amount of iron oxide present in the slag. The color changes from black to dark brown-green, to light brown, to white, and then to gray. Both the gray and white slags are disintegrating slags and break down into powders. These last named slags are low in FeO, never having more than 1.00 per cent FeO present, while the average FeO content is about 0.75 per cent.

The light-brown slags usually give off a faint odor of carbide upon being quenched in water. They contain about 0.85 to 1.25 per cent FeO and a CaC_2 content of 0.50 to 0.75 per cent. The black oxidizing slags will contain from 15 to 25 per cent FeO, depending on iron-ore additions.

From present available information the viscosimeter test appears not to have been used with the basic-electric slags. This is because fluorspar is used rather extensively in preparing the refining slag, rendering the viscosimeter useless. If the oxidizing slag is removed from the bath, the refining slag-forming materials will usually contain fluorspar in varying quantities. If the oxidizing slag is not removed, fluorspar is added as the FeO content of the slag drops in order to keep the slag from becoming too viscous. Fluorspar is added in preference to sand with the purpose of maintaining a high basicity. Ratios of lime to silica of 1.5:1 to those of 4:1 have been used. Studies on these variations show that there is little change in their abilities of reducing FeO. The quickest fluxing slag is one with a lime-silica ratio of 2.5 to 1. A slag of this ratio with a little fluorspar present is the easiest and quickest fluxing slag that can be formed and is therefore the one frequently used.

Lime-silica ratios can be estimated from the appearance of slag pancake samples, as practiced in the basic open-hearth process. Also, some operators are able to estimate the FeO content of the slag by observing films on the top surface of the pancake test. Three main types of slag are recognized:

1. Basic slag. The center of the pancake sample is a shiny black, with edges of a silvery luster. The bottom of the sample has a dull luster. The lime-silica ratio is about 3.0.

2. Intermediate slag. The top surface of the sample is shiny black. The bottom is dull to shiny black. Lime-silica ratio is from 2.2 to about 2.8.

3. Basic-acid slag. The top surface of the sample is covered with furrows. The bottom is shiny. The lime-silica ratio is about 1.5 to 2.0. If the sample shows a very rough top surface, creased all over, the lime-silica ratio is approximately 1.1 to 1.5.

Single-slag Process.—The single-slag process is used in many cases when the phosphorus content of the scrap is less than 0.04 per cent. One foundry reports that the majority of their scrap contains less than 0.035 per cent phosphorus, and hence the single-slag operation is always used.

The single-slag procedure, of course, speeds up the practice since the deoxidation or refining period can proceed as soon as the charge is melted. This provides a production time similar to that of the acid-electric practice, which is a single-slag process, with the added feature that the sulphur content may be materially reduced.

The charging and melting methods are identical to those already described under the double-slag process except that, by the time melting is complete, a slag of the proper volume and consistency for the refining conditions has been produced.

As soon as melting is complete, the bath is recarburized with a low-phosphorus pig iron, and pulverized-carbon additions are made to the slag, as has previously been discussed under the double-slag process. Since there are no outstanding differences in the refining practice of the two operations, it is reasonable to expect the same degree of desulphurization in the single-slag method as was obtained in the double-slag operation. There is no phosphorus elimination, of course, since all the phosphorus oxide formed during the melting-down stage is reduced under the carbide slag of the refining period.

Typical finishing slags as used in the single-slag method are illustrated in Table XIII. It will be noted that slag No. 5 is high in iron oxide. Some producers maintain that this is necessary for a steel with good fluidity requirements, since it is reasoned that, because of the equilibrium conditions between the metal and the slag, a higher FeO content in the slag will produce a higher FeO content in the metal and hence a greater

fluidity is obtained.   This is true, provided that a sufficient length of time is allowed for the equilibrium to take place.   In the short period of 30 to 45 min. in which the refining slag is active, the equilibrium condi-

TABLE XV.—RECORD OF A SINGLE-SLAG* HEAT OF A 3-TON FURNACE

Analysis desired: C, 0.25 per cent; Mn, 0.60 per cent; Si, 0.30 per cent; P and S, less than 0.04 per cent

Charge:

	Lb.
Miscellaneous scrap low phosphorus	5,000
Foundry returns	3,000
	8,000

Time	Activity
3:30 P.M.	Power on high voltage
4:00	Lime 50 lb.
4:30	Lime 50 lb.
5:00	Lime 75 lb.
5:30	Lime 50 lb.
5:45	Melted down—first laboratory test for C, Mn, and P
5:48	Slag test No. 1
5:50	Add: ferromanganese 40 lb., 80 per cent Mn
5:55	Add: wash metal 230 lb., 3.50 per cent C
5:58	Add: pulverized carbon 10 lb., low voltage
6:00	Report first test C 0.12 per cent, Mn 0.13 per cent, P 0.038 per cent
6:05	Add: fluorspar 25 lb.
	pulverized carbon 10 lb.
6:13	Slag test, No. 2, faint acetylene odor
6:15	Second laboratory test for C and Mn
6:15	Pulverized carbon 5 lb.
6:20	Slag test, No. 3 strong odor of acetylene
6:25	Pulverized carbon 5 lb.
6:25	Temperature test, intermediate voltage
6:27	Slag crumbles to white powder
6:30	Report second test: C, 0.20 per cent, Mn, 0.45 per cent
6:30	Add: wash metal 90 lb., 3.50 per cent C
6:35	Add: ferromanganese 11 lb., 80 per cent Mn
	Add: ferrosilicon 60 lb., 50 per cent Si
6:40	Stir bath; temperature test
6:42	Slag test No. 4
6:43	Tapped

Ladle analysis: C, 0.25 per cent; Mn, 0.62 per cent; Si, 0.30 per cent; P, 0.035 per cent; S, 0.020 per cent.

* The corresponding slag analyses for the single-slag heat are given in Table XVI.

tions are seldom if ever reached, and therefore a slag containing 0.75 per cent FeO may show as much FeO in the metal as a slag containing 2.50 per cent FeO.   This, of course, depends on the rate with which FeO is reduced from the slag.   It also should be pointed out that the FeO content (in the metal) at which maximum fluidity is obtained is not as yet

known. It is, however, known that too much FeO in the metal is not conducive to good fluidity.

A furnace chart of a commercial single-slag heat is presented in Table XV for study and comparison with the double-slag practice of Table XII.

The slag analyses as presented in Table XVI clearly point out the trend in the return of the phosphorus to the metal. It also shows the drop that takes place in the CaF₂ content after the single addition of

TABLE XVI.—SLAG ANALYSES FOR SINGLE-SLAG HEAT OF TABLE XV

	Oxidizing slag 5:48 P.M. No. 1	Refining slag		
		6:13 P.M. No. 2	6:20 P.M. No. 3	6:42 P.M. No. 4
CaO	51.20	46.09	49.61	53.55
CaF₂	0	7.39	4.46	3.40
CaS	0.32	0.41	0.44	0.52
CaC₂	0.09	3.13	4.62	2.88
FeO	14.85	3.02	1.02	0.56
MnO	5.52	3.53	1.31	0.61
Al₂O₃	2.94	2.89	3.00	2.82
MgO	7.76	8.40	8.62	9.09
Cr₂O₃	0.57	0.16	0.06	0.02
P₂O₅	0.47	0.33	0.16	0.07
SiO₂	15.39	25.86	26.35	26.60
Color	Black	Brown	Gray	White

fluorspar early in the refining part of the heat. An indication of the amount of phosphorus that is eliminated in a single-slag heat may be obtained from the phosphorus analyses of the heat recorded in Table XV. A sample taken just at the time of melt-down showed 0.038 per cent phosphorus, whereas the ladle analysis gave 0.035 per cent phosphorus, or a reduction of only 0.003 per cent.

Alumina-slag Modification of the Double-slag Process.—A modification of the double-slag process has been devised in which aluminum is used in place of carbon to obtain a reducing slag. This type of slag finds its greatest application in stainless steel or other alloys where it is desired to maintain the carbon content in the metal at a low figure. The method has equal application to carbon steels and low-alloy steels.

The melting and oxidizing stages are similar to those used in the double-slag or single-slag practice. A furnace record using aluminum additions to the refining slag is given in Table XVII.

There are certain unusual features exhibited by this practice, one of which is the addition of ferrosilicon before the first slag is removed. The reason given for this is that slag can be removed more easily and more

completely from a killed heat. From the standpoint of phosphorus removal, it would seem to be better to add it after the slag has been removed. The large amount of ferrosilicon and ferromanganese added during this period was for the purpose of killing the bath. The refining is then carried out on a killed bath.

TABLE XVII.—FURNACE RECORD DOUBLE-SLAG HEAT ALUMINA-SLAG MODIFICATION
Analysis desired: C, 0.30 to 0.35 per cent; Mn, 0.55 to 0.65 per cent; Si, 0.30 to 0.40 per cent; P and S, less than 0.03 per cent

Charge:

	Lb.	Per cent Mo
Molybdenum foundry scrap	9,000	50
Plate scrap	7,000	
Calcium molybdenum	84	40
Limestone	400	

Time	Activity
1:00 P.M.	Current on
2:00	Lime 200 lb. added
2:45	Iron ore 150 lb. added
2:50	Heat melted
2:55	Laboratory test for carbon: reported C, 0.15 per cent
3:15	Laboratory test for Ni and Mo: reported Ni, 0.15 per cent, Mo, 0.56 per cent
3:30	Add: FeSi 40 lb., Si, 50 per cent
3:35	Slag off
3:38	To open bath add: FeMn 40 lb., Mn, 80 per cent; FeSi 40 lb.; coal 12 lb.
3:41	Slag materials mixed and added: lime 400 lb., fluorspar 75 lb., ground coke 20 lb.
3:45–4:25	Add: 200 lb. lime and 65 lb. aluminum shavings
3:55	Laboratory test for C and Mn: reported C, 0.21 per cent, Mn, 0.38 per cent
3:55	Slag test No. 1
4:10	Add: 24 lb. FeMn and 275 lb. pig iron
4:15	Add: 53 lb. FeSi (50 per cent Si)
4:25	Add: 20 lb. FeTi (40 per cent Ti)
4:30	Slag test No. 2
4:35	Tapped

Ladle analysis: C, 0.32 per cent; Mn, 0.57 per cent; Si, 0.37 per cent; P, 0.024 per cent; S, 0.017 per cent; Mo, 0.60 per cent; Ni, 0.16 per cent.

Mechanical properties:
Annealed at 1650°F.

Tensile strength	77,750 p.s.i.
Yield strength	45,750 p.s.i.
Reduction of area	45 per cent
Elongation	27.5 per cent
Bend	120 deg.

TABLE XVII.—FURNACE RECORD DOUBLE-SLAG HEAT ALUMINA-SLAG
MODIFICATION.—(*Continued*)

	Refining slag	
	3:55 P.M. No. 1	4:30 P.M. No. 2
CaO	55.70	46.44
CaF₂	9.06	4.86
CaS	0.39	0.57
CaC₂	1.66	0.22
FeO	0.62	0.51
MnO	0.77	0.34
Al₂O₃	8.90	12.24
MgO	8.69	15.84
SiO₂	12.62	18.34
TiO₂	....	0.86
Color	Green to white	White

Lime should be mixed with the aluminum so that it may be spread more evenly over the bath and so that it may prevent the slag from becoming too fluid. Analyses of the metal have shown that there is no aluminum pickup. The steel foundry regularly using this slag states that the metal can be shanked if necessary.

Alloy-steel Production.—There is very little difference in the process between carbon-steel practice and low-alloy steels. In the case of nickel, copper, or molybdenum steels, a large portion of the nickel, copper, and molybdenum in the form of calcium molybdenum is charged with the scrap. Melting is carried on as rapidly as possible and from time to time lime or limestone is added to make up the slag to the proper consistency. If iron ore is needed, it is added at the time that the scrap is completely melted. As soon as the boiling action from the ore has subsided, the oxidizing slag is removed. The usual refining slag is placed on the bath, and reducing conditions are maintained. Alloys such as chromium and manganese are not added until a carbide slag is obtained. In case of large additions of alloys, care should be taken not to add too much at one time, or the bath will be chilled and extra time will be required for melting. After addition of alloys, the bath should be thoroughly stirred before samples are taken for analyses. The furnace record of a typical heat of manganese-molybdenum alloy cast steel is shown in Table XVIII.

Stainless steel is seldom, if ever, made under a carbide slag, and if the alumina slag is not used, pulverized ferrosilicon is added to the slag to bring about reducing conditions. If stainless-steel scrap is melted, only the single-slag process is used.

High-manganese steel (12 to 14 per cent) is always produced in the basic-electric furnace. The furnace bottom is composed of magnesite

that is sintered in place. Only the single-slag process is used in the production of 12 to 14 per cent manganese steel. A definite melting procedure based mainly on the power input is followed for all heats. The melting time will average about 60 per cent of the total time for producing a heat. Lime is added to the bath during the melting period.

TABLE XVIII.—FURNACE RECORD OF MANGANESE-MOLYBDENUM ALLOY CAST-STEEL HEAT

Analysis desired: C, 0.25 to 0.30 per cent; Mn, 1.40 to 1.55 per cent; Si, 0.35 to 0.45 per cent; P and S, 0.035 per cent max.

Charge:

	Lb.
Mn-Mo foundry returns	8,000
Forging scrap	8,000
	16,000
Mill scale	400

Time	Activity
7:45	Power on
9:15	Lime 100 lb., fluorspar 25 lb. added
9:45	Lime 100 lb., fluorspar 25 lb. added
10:20	Melted down, first test: C, 0.15 per cent; Mn, 0.25 per cent
10:25	Slag off
10:32	FeSi, 50 per cent, 16 lb. (0.05 per cent Si)
10:34	FeMn, 80 per cent, 220 lb. (1.10 per cent Mn)
10:34	Molybdic oxide briquettes 8, 20 lb. Mo (0.12 per cent Mo)
10:35	Second slag: lime 400 lb.; fluorspar 100 lb.; pulverized coke 30 lb.
11:00–11:35	Lime 100 lb.; and pulverized coke 70 lb.
11:00	Pig iron 188 lb. (0.05 per cent C)
11:05	Second test: C, 0.24 per cent; Mn, 1.25 per cent; slag foamy and brown
11:20	Slag gray and powdering
11:30	FeSi 128 lb. (0.40 per cent Si); FeMn 30 lb. (0.15 per cent Mn)
11:35	Pig iron 113 lb. (0.03 per cent C); FeMn 16 lb. (0.08 per cent Mn)
11:40	Al 8 lb. (1 lb. per ton)
11:45	Tapped (slag gray and powdering)

Ladle analysis: C, 0.25 per cent; Mn, 1.46 per cent; Si, 0.39 per cent; P, 0.020 per cent; S, 0.024 per cent.

After melting, the bath will contain approximately 0.90 to 1.00 per cent carbon, 11 per cent manganese, and 0.30 per cent silicon. If a good basic slag is maintained, it will consist approximately of CaO, 50 per cent; SiO_2, 20 per cent; FeO, 3 per cent; and MnO, 10 to 15 per cent. In some cases, very little lime is added to the bath during the melt-down period. In this case a slag is obtained approximately as follows: CaO, 10 per cent; SiO_2, 30 per cent; FeO, 3 per cent; and MnO, 40 per cent

As soon as melting is complete, the slag is deoxidized with pulverized coke. Spar is generally used for shaping the slag to the proper consistency if it is too viscous. A strong carbide slag is not often desired. A typical finishing slag under the above processing is as follows: CaO, 58.66 per cent; SiO$_2$, 24.01 per cent; FeO, 9.42 per cent; MnO, 1.01 per cent. Ferrosilicon is used as a deoxidizer.

In another method of making high-manganese steel, neither carbonaceous material nor pulverized ferrosilicon is used to deoxidize the slag. It is claimed that the amount of manganese recovered by following such a practice will not pay for the time or material used. Also, the mechanical properties do not appear to be improved by the reduction process. A finishing slag for such a practice would consist of CaO, 12 per cent; SiO$_2$, 30 per cent; MnO, 30 per cent; MgO, 20 per cent; and Al$_2$O$_3$, 2 per cent. Metal produced under a slag of this type can be shanked over the entire shop.

Deoxidizers Used in the Basic Process.—Basic-electric steel is treated with ferrosilicon and ferromanganese. These additions also deoxidize small amounts of FeO and MnO in the metal. The alloys are usually added from 5 to 15 min. before tapping. In some cases ferrotitanium is added to the bath about 5 min. before tapping. An amount of 0.05 per cent titanium constitutes the normal addition. Titanium appears to have excellent deoxidizing powers and is a stronger deoxidizer than manganese or silicon but not so strong as aluminum.

Aluminum is used in some cases as a final deoxidizer, the addition being made to the ladle. Low-ductility troubles arising from aluminum additions are not encountered when the critical amount of 1 lb. per ton is added. This is probably due to the fact that the sulphur content in the basic process is usually 0.02 per cent or below, an amount not sufficient to form an extensive network of sulphide inclusions.

REACTIONS IN THE BASIC PRACTICE

Oxidation Reactions. *Carbon.*—The elimination of carbon from the bath takes place by the reaction of the FeO present in the steel with carbon in accordance with the following reaction:

$$C + FeO = CO + Fe$$
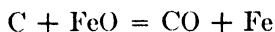
The CO formed escapes as a gas, giving rise to the "carbon boil."

The rate of oxidation of the bath is directly proportional to the oxidizing power of the slag and inversely proportional to the concentration of FeO in the steel. Rapid solution of FeO in the steel gives rise to a more violent carbon boil. The viscosity of the slag exerts considerable influence on the rate of oxidation of the bath by the slag.

The concentration of active iron oxides in the slag is determined by

the carbon content of the bath, where the slag is sufficiently fluid   As the carbon drops, a higher concentration of FeO is necessary to bring about the subsequent drop.   This is due to the fact that equilibrium requires the presence of a higher percentage of FeO in the metal with lower carbon content.   It is generally believed that the silicon and nearly all the manganese and phosphorus are oxidized before the oxidation of carbon becomes active.

Silicon.—The silicon content of the charge is eliminated by oxidation largely by the direct action of the flame during the melting period in the open hearth and by iron oxide introduced with the scrap as rust or scale in the electric and open-hearth furnaces in accordance with the following reaction:

$$Si + 2FeO = SiO_2 + 2Fe$$
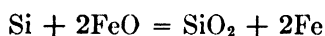
By the time the steel is melted, the silicon content is at the low figure of about 0.04 per cent or less.   Because silicon has a much stronger tendency to react with FeO than carbon, the silicon must be nearly completely eliminated from the bath before any substantial oxidation of carbon can occur.

Manganese.—The reaction for the oxidation of manganese is

$$Mn + FeO = MnO + Fe$$
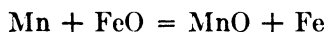
The amount of manganese oxidized from the bath varies with the amount of manganese in the charge, the amount of manganese in the slag, the slag volume, and the carbon content and temperature of the steel at the time the heat is blocked (in the open hearth) or at the time the oxidizing slag is removed (in the electric).

For a single temperature, the lower the value for the ratio of FeO to MnO, the higher will be the manganese residual.   Temperature has a very marked effect upon the manganese-equilibrium constant.   This explains the reason why the residual-manganese content of the bath in the open hearth may be increased during the latter part of the refining period as the bath temperature is increasing, in spite of the fact that the slag becomes more oxidized and the ratio of FeO to MnO shows an increase in value.   The residual-manganese content seldom goes below 0.10 per cent and the normal manganese residual in the open hearth is usually near 0.15 to 0.20 per cent.

Phosphorus.—The reaction for the elimination of phosphorus is written in a number of ways.   The main reaction, put in its simplest, is the oxidation of phosphorus by FeO:

$$2P + 5FeO = 5Fe + P_2O_5$$

Since this reaction is easily reversed and phosphorus pentoxide does not

exist in liquid steel, the reaction is considered to be

$$2P + 5FeO + 3CaO \rightarrow 3CaO \cdot P_2O_5 + 5Fe$$
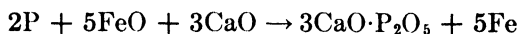
The equilibrium expression from this reaction involves the concentration of tricalcium phosphate and of free lime in the slag, of FeO in either the slag or the bath, and of phosphorus in the bath.

The requirements for the removal of phosphorus from the steel call for both a basic and an oxidizing slag. The oxidizing conditions as signified by the FeO content of the bath and the slag are largely a function of the carbon content of the bath under equilibrium conditions. Also, the oxidizing conditions are controlled by the FeO/MnO ratio in the slag. High MnO contents retard dephosphorization by diminishing the state of oxidation. Also, high MnO contents reduce the lime-carrying power of the slag and also the dephosphorizing power, since the sum

$$CaO + MgO + MnO$$
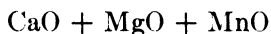
is approximately constant at 61 per cent for medium-carbon steels.

Excessive basicity of the slag hinders the removal of phosphorus by creating a viscous slag and sluggish action. The speed with which phosphorus can be removed from mild steel by basic slags appears to vary with the fourth power of the viscosity of the slag. Temperature has an effect upon phosphorus reactions very similar to its influence on the manganese equilibrium, for higher bath temperatures give a higher residual-phosphorus content.

Sulphur.—The removal of sulphur from steel in the open-hearth practice takes place according to the reaction

$$FeS + CaO = FeO + CaS$$
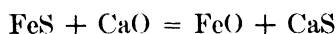
The transference of sulphur from the steel to the slag will depend on the ratio of CaO to FeO and the amount of slag available to act as an absorber. The MnO content of the slag will act equally with FeO in preventing the sulphur transference to the slag. High temperature aids in the removal of sulphur by permitting the use of a slag high in CaO content.

Manganese plays an important part in sulphur removal by the reaction

$$FeS + Mn = MnS + Fe$$
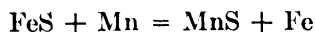
This reaction takes place mostly during the melting period. Sulphur may also be eliminated in the open-hearth practice by the addition of ferromanganese or spiegel to the bath prior to the block or by the blocking action. The removal takes place according to the above reaction.

Reducing Reactions (Basic-electric). *Iron Oxide.*—The reaction for the removal of FeO from the slag by the addition of powdered car-

bonaceous material to the slag is

$$3FeO + CaC_2 \rightarrow CaO + 2CO + 3Fe$$

If aluminum or pulverized ferrosilicon is used as the reducing material, the reaction is

$$3FeO + 2Al \rightarrow Al_2O_3 + 3Fe$$
$$FeO + 2Si \rightarrow SiO_2 + Fe$$
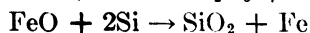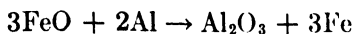
Manganese oxide is reduced in a like manner according to similar reactions. The speed with which these reactions take place depends
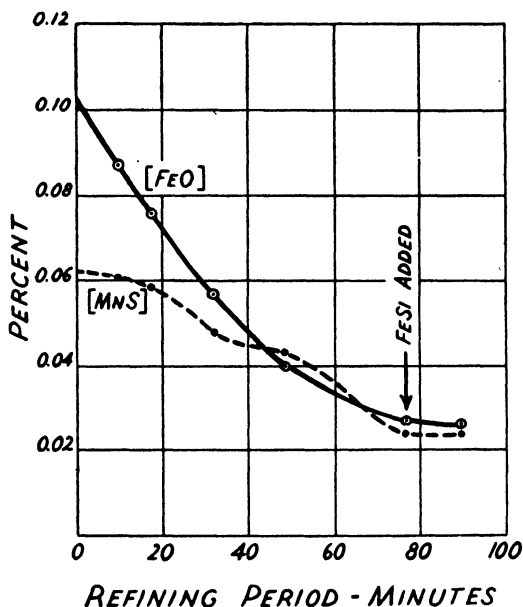
Fig. 14.—Iron oxide and manganese sulphide elimination in a basic-electric heat during refining period. (*Herasymenko and Valenta.*[10])

upon the rapidity with which the dissolved oxides in the bath reach the slag-metal interface. They also depend upon the amount of oxides that are present in the slag. The oxides in the slag soon reach a low figure after the reducing element is added to the slag. Since there is an equilibrium that exists between the ferrous and manganese oxides in the bath and in the metal, there is a migration of FeO and MnO from the metal to the slag in order to restore the equilibrium. A typical curve of iron oxide elimination from the bath under the reducing slag of the basic-electric furnace of 6-ton capacity has been prepared by Herasymenko and Valenta[10] and is shown in Fig. 14. This figure refers to an 0.18 per cent carbon steel that did not require recarburization before the reducing

slag was formed. The diagram shows that the amount of iron oxide in the metal decreases rapidly at first. Toward the end of the heat, the rate of iron oxide elimination becomes slower. This is because the rate of diffusion depends on the concentration difference between the slag and the metal, and the rate of iron oxide elimination will be faster as the total concentration of iron oxide in the bath is increased.

The minimum content of iron oxide that can be obtained in basic-electric steel under a good reducing slag is approximately 0.020 per cent. This value is considerably lower than the amount of dissolved iron oxide in acid steel and thus basic-electric steel can be cleaner.

Sulphur.—There appear to be several factors affecting the removal of sulphur. In order to obtain maximum sulphur elimination, (1) the slag must be highly basic, (2) a reducing slag must be present, (3) the slag must be fluid and hot, and (4) the slag should probably contain additions of fluorspar.

The principal desulphurizing reactions appear to be

$$2CaO + 3MnS + CaC_2 \rightarrow 3CaS + 2CO + 3Mn$$
$$2CaF_2 + 2MnS + Si \rightarrow 2CaS + SiF_4 + 2Mn$$
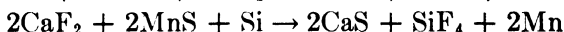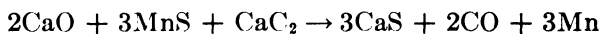
A good refining slag will meet the above requirements for maximum sulphur elimination. Fluorspar is of great assistance in removing sulphur. By increasing the fluidity of the slag and thereby increasing its activity, it speeds up the reactions taking place; it also permits very basic slags to be employed, which would otherwise be too viscous to be worked. There is, furthermore, a possibility that fluorspar is effective in removing sulphur by combining with it, forming a volatile sulphur-fluorine compound that passes out of the furnace as a gas.

A complete review of the physical chemistry of steelmaking and of slags is too extensive to be presented in this book.[1] A discussion of deoxidation is presented also in Chap. III.

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CHAPTER II

ACID PRACTICE FOR THE PRODUCTION OF STEEL FOR STEEL CASTINGS

ACID OPEN-HEARTH PROCESS

There are at this date approximately 130 open-hearth furnaces available for steel melting in the steel casting industry, and of this total 70 are being operated by the acid process producing a normal rated capacity of approximately 24,000 tons per month. This is considerably below the maximum monthly capacity if worked on a 24-hr. basis. The furnaces vary in size from 10 to 125 tons rated capacity and are fired by gas or oil.

The acid open-hearth process is limited generally to the production of dead-killed steel, and practically all acid open-hearth steel produced is used for the manufacture of steel castings or ingots for forgings. It is generally considered, both in this country and in Europe, that the acid open-hearth steelmaking process is one that has the possibilities of producing a high quality of dead-killed steel within narrow limits of composition. The process is one that does not permit sulphur or phosphorus to be removed from the metal. The slag consists essentially of silica, iron oxide, and manganese oxide. The iron oxide content of the slag is the most active oxidizing constituent; yet as reported by Herty and Jacobs[13] it is only approximately one-third to one-half as active, comparatively, in the acid as in the basic furnace. This feature of the process is probably the principal reason why the production of high-quality steel can be consistently carried out in the acid open-hearth furnace with a minimum of control features. The physical chemistry of the process is not entirely understood.[43]

A study of the proceedings of the Acid Open-hearth Conferences held by the American Institute of Mining and Metallurgical Engineers will reveal that there are material differences in the acid practice of the various foundries throughout the country. In general, it may be said that such differences are the result of local experiences and needs, and yet all are providing an acceptable steel for casting production. It is not the purpose to present here all the various modifications that may be found in the production of steel by the acid practice but to describe the general features.

Bottom Construction.—The principal desire of most melters is to have a sharp-working furnace. By "sharp working" the melters refer to a fast-melting furnace with an abundant supply of good fuel and gas developed by well-designed burners and an efficient regenerator system. In a sharp-working furnace the melter has adequate control of the bath temperature. Slow-working furnaces often require various modifications of melting procedures, and considerable skill is required to prevent cold-melting and the effects accompanying it. Adequate control of temperature is perhaps the most important single factor associated with the process.

A brief discussion of open-hearth furnace construction is given in Chap. I. The primary difference between the basic and the acid furnaces lies in the fact that the walls and roof of the latter are lined with silica brick.

The usual procedure in making a bottom for the acid furnace is to use a good grade of silica sand and to build it up from the brick lying on the pan to a depth of from 10 to 18 in. Silica brick are laid on the pan to a depth of from 15 to 22 in. One foundry reports the use of a deep layer of silica brick laid on the pan but with only approximately 6 in. of sand built up on the brick.

From the time the furnace has started to warm up, it takes approximately 7 to 10 days to put in a good bottom. The furnace is first dried with a wood fire for about 1 day; then the fuel is turned on, and the burners are reversed about every half hour until the furnace is hot (2850 to 2900°F., about 4 days). Sand is then added at intervals of 20 to 30 min. until a bottom of the necessary depth is obtained. The bottom is sloped from all directions toward the taphole.

About 90 per cent of the acid open hearths use one grade of sand—an Ohio open-hearth silica sand. The silica content of the sand is approximately 96 per cent, and usually the sand is purchased in the graded condition with about 25 per cent of the sand remaining on the 20-mesh sieve. It has been stated[3] that the purpose of this grading is that the coarse particles may be distributed throughout the sand and may enable the heat to penetrate the sand more readily, thereby allowing the entire sand layer to fuse instead of merely the top fraction of the layer.

From available information it seems that a sand that fuses at about 2900°F. is a proper sand for bottom building. This type of sand will fuse on its top surface $\frac{1}{8}$ in. in 5 or 6 min. Fusion will be fairly uniform throughout the added layer in $\frac{1}{2}$ hr. if the graded type is used. Two foundries report that their practice is to put in a good bottom of 10 to 12 in. in 4 to 5 days after the furnace is hot. If the furnace is kept hot, sand may be added at the rate of 700 lb. approximately every half hour. This rate is not kept up for the 4 to 5 days because there are other time-con-

suming duties that require attention, such as making the taphole, fixing ports, etc. For a good bottom a 60-ton furnace will require approximately 55,000 lb. of sand. The bottom sand is added by shoveling.

Bottom building, by making small frequent additions of sand, is also an acceptable practice. A hard, well-sintered bottom can be put down in this manner at the rate of approximately 1,000 lb. of sand per hour. A bottom 14 to 16 in. in depth, built in 80 hr. for a 70-ton stationary furnace, in one case used a total of 85,000 lb. of sand. An acceptable bottom can be put in a 25-ton furnace with 35,000 lb. of sand, while a 50-ton furnace will require about 55,000 lb. and a 100-ton furnace will need approximately 95,000 lb. of sand.

Some foundries use a mixture of fine sand and acid slag in the ratio of one part of slag to nine parts of sand for building bottoms. Others have found it expedient to add a small percentage of fire clay or to make additions of a low-fusion sand to the regular grade of high-fusion fine sand. These modifications are made for the purpose of reducing the fusion temperature of the bottom material so that it will fuse readily. It is believed, however, that these modifications do not produce the hard fused bottom that is possible when a high-quality bottom sand is used alone.

An interesting modification in building a bottom is the reported use of minus 40-mesh powdered 50 per cent ferrosilicon, mixed with sand in the ratio of about one part ferrosilicon to 100 parts of sand. The bottom prepared in this manner has stood up exceptionally well for 3 yr. A few foundries are using a rammed-in bottom composed of the regular acid open-hearth sand and molasses. Only sufficient molasses is used to bond the sand. The bottoms are rammed in cold. The use of rammed bottoms shortens the time for putting in a bottom by 24 hr.

After a bottom is completely built up, it is the practice in some foundries to add acid slag to the furnace, melt it, and splash the bottom with a good coating. This is done so that the slag will penetrate the sand and will not allow the steel melted on the first heat to penetrate easily.

Another foundry reports the use of quartzite blocks on the end slopes above the slag line. The blocks are approximately 4 by 4 by 10 in., and they are set with expansion joints provided. Some spalling is recorded but not sufficient to be of concern. The slopes are maintained very well and thus they produce a constant throat area.

Some of the furnaces have insulated bottoms. In these furnaces a combination of insulating concrete and insulating brick is used. The concrete, 3 in. in thickness, is placed on the flat of the pan, and $2\frac{1}{2}$ in. of insulating brick are placed on the sides and ends of the pan to slag-line height.

There are a number of ways for constructing the taphole. One method is to put into the furnace a water-cooled tube and to build the bottom over it. If a 6-in. taphole is desired, a 4-in. tube is used, since the sand does not sinter all the way to the water-cooled tube. Another method is to use specially shaped taphole brick. Eight bricks shape the circle, and two sets are required. Still another method for shaping the hole is to use an electrode and build the bottom around it. The sand fuses to the electrode. After the bottom is completed, the electrode is rammed out. Wet sacks are rammed into place and faced up as a substitute to the water-cooled tube method. During the building of the bottom it is necessary to open the taphole and reload it with another set of wet sacks. The most commonly used method of taphole construction is to make a wooden plug the size of the hole desired. The plug is covered with a refractory material. By the time the plug is burned out, the taphole is constructed.

Tapholes are generally plugged with coal or with a mixture of buckwheat coal and sand or of coke breeze and sand. The coke or coal content will vary from 50 to 70 per cent of the mix, depending on the desired hardness of the tap plug. This mixture is backed up on both the outside and inside with molding sand.

Some steels such as the low-carbon steels are quite hard on furnace bottoms; hence bottom replacement will depend on the type of steel made as well as the construction of the furnace. A steelmaker who produces nothing lower than 0.50 per cent carbon steel may make a bottom last from 3 to 5 yr. In most steel foundries, furnace bottoms are renewed at the time of every campaign or every other campaign. A campaign comprises the length of time prior to the relining of a furnace. One foundry reports that the bottom is renewed on a 10-ton furnace every campaign, which includes from 1,700 to 2,000 heats.

The hearth sand consumption per ton of steel produced will vary, depending on the type of steel made and the construction of the bottom. Values from 250 to 125 lb. per ton have been recorded. Most of the sand that is added between heats is placed just below the slag line to take care of high-oxide slags. The foundry reporting the highest figure has found that the use of a very pure silica sand, 99.5 per cent, has improved bottom practice when using a low-silicon, low-manganese charge. No attempt is made to sinter this sand in the furnace. Approximately 40 lb. per ton of charge are added immediately before starting to charge.

Holes in the bottom of the furnace are repaired by blowing out the slag with steam, adding sand, and allowing it to sinter. A very good practice is to add a small quantity of crushed ferrosilicon to the hole, deoxidize the FeO in the slag, and then add sand, puddle, and allow it to fuse. It

is believed that FeO is the most active compound in the deterioration of the acid bottom.

Compositions of the working hearth of an acid open-hearth furnace have been reported by Ferguson[1] and are shown in Table XIX. Samples were taken by driving a 1½-in.-diameter tube about 2 in. into a midway position in the banks, approximately 6 ft. from the taphole of a 60-ton furnace. These figures reveal that an iron oxide network, which

TABLE XIX.—COMPOSITION OF THE WORKING HEARTH (60-TON ACID FURNACE)
Ferguson[1]

	Per cent			
SiO_2	70.54	69.24	67.40	73.60
FeO	15.84	18.00	20.40	17.50
Fe_2O_3	5.51	4.44	3.96	3.44
Al_2O_3	2.29	2.65	1.06	2.06
MnO	5.28	4.05	5.02	2.27
CaO	0.08	1.04	2.10	0.84
MgO	0.72	0.86	0.44	0.21
	100.26	100.28	100.38	99.92
Bottom in service	12 weeks	13 weeks	8 years	8 years

is presumably required to cement the silica crystals, gradually corrodes and ultimately fluxes away the silica grains, thus requiring constant renewal of the working face.

The Charge.—The acid open-hearth charge will usually consist of pig iron, foundry scrap (gates and risers), and purchased scrap. The purchased scrap should be of the solid type and as free as possible from hydrated rust. Wet rust is regarded as a source of hydrogen content in the steel and therefore must be kept out as much as possible. During the Second World War, almost any and all scrap was used, including very light sheet scrap, automobile graveyard scrap, and detinned tin cans. Much of this scrap is quite heavily covered with rust. The scrap should have a low content of sulphur and phosphorus—if possible, below 0.04 per cent each—since neither of these elements can be eliminated in the acid process.

The pig-iron content of the charge will vary from 10 to 50 per cent. The average charge should contain approximately 15 to 20 per cent pig iron. The percentage of pig iron is figured on the basis of the carbon content of the charge and its relation to the finishing carbon. Some foundries plan on having the total carbon content of the charge greater than the finishing carbon by about 0.50 per cent, while others believe that a better figure is 0.70 per cent. In some cases even higher values

are desired. Pig iron is added to produce sufficient carbon in the heat at melt-down so that a vigorous boil will result on the addition of iron ore.

It is possible to add carbonaceous material to bring up the carbon content of the charge. Some operators prefer the pig-iron addition in preference to carbonaceous material, especially those operators who desire a high manganese content in the charge. Under normal conditions it is possible to attain this condition by the use of pig iron containing a rather high manganese content.

In the past, most low-phosphorus pig iron as used in the acid open-hearth would average about 1.25 per cent manganese, but during the Second World War very little pig iron was produced containing over 1.00 per cent manganese. It is the usual procedure for foundries to purchase pig iron with a 1.00 per cent maximum manganese content; approximately 85 per cent of all acid open-hearth foundries use this type of pig iron. A few foundries like a high-manganese residual and purchase pig iron, when available, to specifications calling for 1.50 to 2.00 per cent manganese.

Phosphorus and sulphur specifications of pig iron are usually set at 0.035 per cent each, and pig iron purchased under these specifications will record 0.025 to 0.029 per cent phosphorus and 0.025 to 0.028 per cent sulphur. The silicon content will usually be 1.50 to 1.75 per cent.

The average silicon and manganese content of the charge has a considerable influence on the subsequent operations of the bath. The silicon content of the charge will usually be about 0.60 to 1.00 per cent. In some gas-fired furnaces silicon as low as 0.40 per cent in the charge has been reported. These values are believed necessary to prevent overoxidation of the bath and reduce to a minimum the attack on the siliceous banks and bottom of the metallic oxides formed during melting. The actual content of silicon figured for the charge will depend on the silicon content obtained at the time the charge is completely melted.

The manganese content of the charge varies over wide limits of from 0.4 to 1.75 per cent. A number of foundries desire to have the higher values of 1.25 to 1.75 per cent. This often requires large amounts of pig iron or the addition of ferromanganese or spiegel. Most foundries plan on having approximately 0.75 to 1.00 per cent manganese in the charge. Manganese ore can also be used to increase the MnO content of the slag.

The carbon content of the charge will also vary somewhat. If the finishing carbon is to be 0.30 per cent, then the charge should contain about 1.00 per cent carbon.

There are two methods of charging furnaces; one is called "unit charging" and the other "level charging." In unit charging the entire charge is placed in the furnace at one time. In support of this method it is pointed out that intermittent heating and cooling of the furnace are harmful to the furnace refractories and that by charging the heat in

various parts the slag line is cut at different heights. Also, the second and third charges are sometimes put in before the furnace is hot enough, and this condition results in an unsatisfactory melt-down.

Other operators believe that level charging, or charging the heat in two or three parts, is essential for quick melting, rapid slag formation, the avoidance of sulphur pickup, and reducing oxidation. There is ample room above the charge to burn the oil in the level-charging method. It appears that a unit charge will melt faster in a slow-working furnace, whereas a charge is melted more rapidly in a sharp-working, or fast-melting furnace if level charging is used.

A number of practices are followed with respect to the position of the pig iron in the charge. One method is to put all the pig iron on the bottom of the charge. In this way, the pig iron will prevent the oxides of the scrap from damaging the hearth during the melting. In other cases, the pig iron is spread uniformly on top of the charge with the idea of protecting the furnace banks and ends. Two foundries report that the pig iron is placed above a layer of light open scrap, such as springs or turnings, so that the heat will penetrate throughout the bottom of the charge.

The pig iron may be added as a second part of a leveling charge, or it may be added to the bottom, middle, and top of a unit charge. As has been stated, it does not matter where it is placed because it melts first and runs down and protects the bottom. It is believed that the major percentage of the pig iron should be on the bottom of the charge, with the remainder at the top along the banks. It has been suggested that the pig iron on the bottom should not exceed 12 per cent of the charge. In some cases, ferrosilicon is placed along the banks for their protection.

The charging time for a furnace will depend on the character of the scrap and the amount of the charge (size of the furnace). For example, a 145,000-lb. heat can be charged in approximately $2\frac{1}{2}$ hr. total charging time, by a charging machine. The charge is added in two parts. The temperature of the charge averages about 2300°F. when the second part is added.

The Melt.—As soon as the furnace has been charged, the fuel, which was partly turned off during the charging, is turned on full with the flame reversed about every 15 to 20 min. The flame during the melting period is a sharp oxidizing flame with drafts adjusted so as not to obtain too much of a carbon drop during melting. The time required to melt a 145,000-lb. heat after charging is complete is about 3 hr. A 22,000-lb. charge can be melted in approximately 1 hr. in a 10-ton furnace.

The function of the silicon in the charge is to protect the carbon from oxidizing too rapidly during the melting, as well as to prevent erosion of the banks. The manganese in the charge aids in obtaining a comparatively clean bath at melt-down.

At the time the bath is completely melted, the carbon content for casting heats generally is from 0.20 to 0.30 per cent above the finishing carbon content. Thus for a 0.30 per cent carbon steel the carbon content at melt-down will be about 0.50 to 0.60 per cent. During the oxidation period this carbon will be reduced to a value of about 0.15 to 0.20 per cent, and the finishing additions will bring the finishing carbon back to the 0.30 per cent figure. The carbon content at the melt-down should be high enough so that an extensive vigorous boil will be obtained.

The silicon content at the melt-down should not be over 0.20 per cent. If the silicon content is much above this figure, considerable time will be lost in oxidizing the silicon, and the boil will not be vigorous until the silicon content has dropped to below the 0.20 per cent value. If the silicon content is too low, the carbon content at melt-down must be higher than the suggested 0.30 per cent above the finishing carbon, since the rate of carbon drop will otherwise be too fast. A carbon drop at the rate of 60 points per hour is about the safe limit. For best results it is believed that the silicon content at melt-down should be about 0.15 per cent.

The silicon content of the bath at the time of melt-down varies directly according to conditions of the charge, in the following order of probable importance:

1. Silicon content of the charge
2. Working speed or sharpness of the furnace
3. Compactness of the charge
4. Manganese content of the charge

The manganese content of the bath at the time of melt-down also varies directly in the following probable order:

1. Manganese content of the charge
2. Silicon content of the charge
3. Sharpness of the furnace
4. Compactness of the charge

With these points in mind the manganese content will vary as follows:

Manganese Content of Charge, Per Cent	Manganese Content at Melt-down, Per Cent
1.25–1.75	0.30–0.40
0.60–1.00	0.15–0.20

From the above it may be seen that the amount of silicon in the charge has more influence on the amount of manganese present at melt-down than the manganese has on the silicon content.

The compactness of the charge is an important determining factor on the silicon and manganese contents at melt-down, owing to the fact that the smaller the surface area per pound of charge, the smaller will be the proportion of iron oxide formed during the melting. This point should be carefully considered, especially when changing from scrap

consisting largely of crops, wheels, couplers, knuckles, and the like to light scrap composed of turnings, auto-body stock and frames, or sheet punchings.

The manganese and silicon contents at melt-down will also depend upon the type of furnace and the fuel and atomizing method used. Sharp-working furnaces and those that permit heats to be shaped up quickly can use lower carbon, manganese, and silicon contents at melt-down. There has been considerable discussion over the advantages to be derived by using a high manganese content in the charge and at melt-down. Certain acid open-hearth operators have stated that a heat will work more satisfactorily when it melts with a manganese content at 0.30 to 0.40 per cent than at 0.15 to 0.20 per cent. With the higher manganese content more fluid-working slags are produced with a lower FeO content, since the MnO replaces some of the FeO. It has also been stated that cleaner steel is obtained with improved mechanical properties. High MnO slags also permit closer control of the carbon drop during the working and finishing of the heat.

The high manganese content of the charge permits a MnO content of 25 to 30 per cent in the slag. The MnO is largely in a state of combination with silica,[2] which allows some of the iron oxide to be freed from its combination with silica. The released FeO diffuses into the metal. Under these conditions the iron oxide of the slag would be about 12 to 15 per cent. With iron oxide at 12 per cent, without the high MnO content, the slag will be very thick, heavy, and gummy. With the high MnO content the slag will be thin, but not watery, if the FeO content is only 12 per cent. Additions will go through the slag easily; the heat works fast and picks up temperature fast.

It has been stated[4] that with the high-manganese charge less erosion of the furnace lining is obtained. Furthermore, there is present more manganese to combine with sulphur to form manganese sulphide inclusions, which are less harmful to the steel than the iron sulphide type of inclusion. The use of the high manganese in the charge requires the use of more iron ore to bring the carbon down to 0.15 to 0.20 than required when the charge contains manganese between 0.75 to 1.00 per cent. Typical compositions of the slag at melt-down are as follows:

Charge type	SiO_2, per cent	FeO, per cent	MnO, per cent
Low manganese................................	50	35	10
High manganese................................	50	20	25

The Oxidation Period.—A good, vigorous boil is desired by all casting producers using the acid open-hearth process. A long vigorous boil is

obtained by the addition of iron ore to the bath to bring about the oxidation of carbon in the melt, producing CO.   The bubbles of CO gas rising out of the bath give the appearance of boiling.

Since the oxidation of carbon to carbon monoxide is endothermic, a high temperature is required before a vigorous boil develops.   Studies made by the use of the carbon-silicon carbide thermocouple[3] showed that a temperature of 2750°F. should be reached before the addition of ore if the proper boil is to be obtained.

It is the practice in most foundries to add the iron ore to the bath almost as soon as the bath is melted, and not wait until the bath reaches a certain minimum temperature.   This practice has largely resulted from using charges that are very similar in character day after day, so that the amount of iron ore needed is fairly well stabilized.   It must be remembered, however, that present-day charges may change in character rather frequently, so that set rule-of-thumb practices may not be sufficient to produce quality steel.   The loss in carbon is slight until the proper temperature is obtained for a vigorous boil.

The use of large quantities of iron ore causes a very fast carbon drop at first with the result that there is a greater absorption of heat due to the oxidation of the carbon.   The boil then falls off in vigor because of insufficient heat in the bath.   The slag thins out because of the higher FeO content and attacks the hearth.   If the ore additions are too light, the speed of reactions is unduly slowed down, the FeO content of the slag is lowered, and the slag becomes thick and viscous.

The boil should be vigorous during the first part of the oxidizing period and should taper off during the latter part, the rate of carbon drop of the initial period being probably twice as great as at the end.   Such conditions permit a better grade of steel.

The rate of carbon drop will vary somewhat, depending on the construction of the bath.   In shallow baths a rate of carbon drop of from 0.40 to 0.55 per cent per hour in the initial hour may be attained.   It is believed that the rate of carbon drop at tapping is generally in the neighborhood of 0.12 to 0.20 per cent per hour.   This rate, at which the carbon drops, is not too important from the standpoint of producing quality steel. It is more important to obtain the desired slag condition, the proper temperatures, and low-silicon residual.

A vigorous boil during the early stages after melt-down is necessary for the elimination of gases and inclusions and is a requisite for the production of quality steel; any carbon drop that will yield these conditions is sufficient regardless of the rate of drop.   It is furthermore believed that an active bath should be maintained at all times to prevent the accumulation of gases.

Some foundries have a rule of not permitting further additions of iron

ore after 1 hr. prior to the blocking of the heat. If further carbon drop is necessary, it is obtained by the use of lime. These steps are taken to permit the FeO content of the slag to drop. There is less chance of obtaining dirty steel if rules similar to these are followed.

Refining.—Refining is generally understood to mean

1. The cessation of ore feeding
2. The slowing down of the reactions of the process
3. The final steps in obtaining a proper slag

The silicon and manganese residuals present after the boil will vary somewhat, depending on the amount of manganese present in the charge and the type of boil obtained. The silicon content will usually be from 0.05 to 0.10 per cent. Most melters believe that the residual silicon should be preferably about 0.06 per cent, although the value of 0.10 is not considered poor practice when the manganese residual is high. American practice requires that at some time during the heat the silicon residual of the bath should be less than 0.07 per cent for the purpose of making certain that any gases picked up during the melting period are eliminated.

The residual manganese will be approximately the following:

Manganese in the Charge, Per Cent	Residual Manganese, Per Cent
0.60–1.00	0.06–0.10
1.25–1.75	0.12–0.20

Many foundries take frequent test samples for carbon analyses, and when the carbon reaches the desired figure, the heat is blocked and tapped. This practice can easily be employed with the use of rapid carbon analyzers.

In a few foundries the carbon content of the bath is allowed to drop to 0.10 to 0.15 per cent. The carbon content is then brought back to the finishing carbon by the addition of recarburizers as well as the deoxidizers. The rate of carbon drop slows up greatly in the neighborhood of 0.10 per cent, especially if the iron oxide in the slag has been reduced as the heat proceeds, which is the usual method. Carbon tests can be estimated quite accurately by visual observation of a bar fracture at the lower carbon contents. There is an advantage in this if the foundry does not have an instrument for the rapid analysis of carbon.

During the period of the carbon drop, some foundries add a small percentage of lime or limestone to the slag. This is done to increase the fluidity of the slag, particularly when the MnO content of the slag is low. Manganese-ore additions will produce a fluid, easily workable, but not highly oxidizing slag.

Lime additions to the slag will also aid in carbon removal but at a slower rate than can be obtained by a same quantity of iron-ore addition. The lime frees some of the iron oxide of the slag from combination with the silica, and the freed iron oxide then reacts to oxidize the carbon of the bath. Lime may be added toward the finish of a heat if there is a need for a faster carbon elimination but not so fast as would be obtained by the addition of iron ore.

The CaO content of the slag will vary, depending on the practice; but it is not necessary that this be greater than 10 per cent. Some finishing slags will show only 3 to 6 per cent CaO. Lime can be added at any time of the heat, and it is ordinarily added after the iron ore has been added, when the boil is active and as the slag is beginning to thicken. Additions of large amounts are made early, such as approximately $1\frac{1}{2}$ hr. before deoxidation. Small quantities of lime are often added within 30 min. of blocking, in order to bring about a continued gentle carbon drop or to correct slag fluidity. Low-carbon heats require lime additions, since there is sufficient FeO present in the slag to maintain fluidity. Lime additions are made to alloy heats in a similar fashion. It has been noted that if too much lime is added, a very watery slag is produced, which is responsible for cutting the banks of the furnace.

Blocking and Deoxidation.—There are two ways of finishing a heat: first, by stopping the carbon drop by means of the addition of deoxidizers to the bath; second, by allowing the oxidation to cease by permitting time to establish equilibrium conditions. In nearly all cases, the former method constitutes foundry practice. The first method requires the use of fluid finishing slags to secure clean steel.[5] Viscous finishing slags prevent the absorption of nonmetallics from the metal and cause a throwback of inclusions in the ladle.

Silicomanganese, spiegel, or ferrosilicon and ferromanganese are used as the blocking agents and as sources of the silicon and manganese necessary to meet specification requirements. Such additions practically stop the further reduction of carbon by the formation of oxides. If the iron oxide content of the bath is low, the silicate inclusions in the metal need not be high.

The use of the first method facilitates production. In the second method, the bath conditions are controlled so that with time the metal and slag reach a passive condition or equilibrium at the end of the heat. At this stage the bath may reduce silicon from the slag. The bath condition is indicated by the cessation of the boiling action and a flat bath.

Silicon Reduction.—It has been stated by certain European operators that the best acid open-hearth heat is produced when silicon is reduced. Foundrymen in this country believe that the second method produces a steel of lower fluidity and that a silicon pickup of more than 0.03 to 0.05

per cent is decidedly detrimental to fluidity. Increased opportunity for blow holes in the solidified steel has also been traced to the silicon pickup process. A high-manganese addition to a low-carbon heat will produce a silicon pickup, but this condition is different since the heat is blocked with the manganese addition.

Silicon reduction from the slag is easier with high bath temperatures and with steel of high carbon content. Silicon reduction can be prevented or kept to a low figure by increasing the draft, which appears to lower the bath temperature. If the FeO content of the slag is low, around 10 to 12 per cent, and the MnO content is low, it is possible to obtain a silicon reduction at the end of a heat just prior to blocking. This is not serious, provided the reduction is small and extends only a brief period.

A partial block some 25 to 40 min. prior to adding the deoxidizers has been employed by using spiegel in quantities of 0.8 to 1.5 per cent of the charge. This partial block is used in sharp-working furnaces where the rate of carbon drop is high. Action in the bath continues after the spiegel addition, but the rate of carbon drop is low.

Silicomanganese, ferromanganese, and ferrosilicon additions are being used by the foundry industry in the acid open-hearth process. Reports from the foundries indicate that the use of silicomanganese does not produce any outstanding beneficial results. A slight saving in the time between making the additions and tapping is brought about by the use of silicomanganese. The heat is tapped 10 to 15 min. after the silicomanganese is added. If the two are added separately as ferroalloys, it is the usual practice to add the ferrosilicon first. The ferromanganese is added about 5 to 8 min. after the ferrosilicon addition, and the heat is tapped 10 to 15 min. later.

Final Additions.—A few operators add the ferromanganese ahead of the ferrosilicon. The heat is tapped as soon as possible after the addition of the ferrosilicon, because of the affinity that silicon has for gases; it is disliked as a "block" for this reason and should not be present in a quiet bath any longer than the minimum time. Accordingly, some prefer spiegel or ferromanganese additions for blocking the heat. The recovery of manganese is often low when ferromanganese is added first. However, this is not the case if the MnO content of the slag is high before the addition of ferromanganese. Since most operators use a low-MnO-content slag, the ferrosilicon is added first, permitting a greater recovery of manganese from the ferromanganese addition.

It has been reported that when the ferromanganese is added after the ferrosilicon, the percentage of manganese recovered is controlled by the length of time that the metal remains in the furnace prior to tapping. Losses under normal conditions range between 30 to 40 per cent of the

amount added if the MnO content of the slag is around 10 to 12 per cent.

Some ferromanganese may be added to the ladle, especially when low-carbon medium-manganese steels are being made. The manganese is added to the furnace in an amount normal for a simple carbon heat, and the additional manganese is added to the ladle. It is believed that 60 to 70 points of manganese may be added to the ladle whenever desired.

Aluminum is added to the ladle in quantities of 0.01 to 0.06 per cent, as a deoxidizer and grain refiner. Not all acid open-hearth heats employ the addition of aluminum to the ladle. Some operators have encountered the low-ductility problem in conjunction with the use of 0.01 to 0.03 per cent aluminum as reported by Sims and Dahle,[6] while others use the critical amounts of 0.02 per cent aluminum and do not obtain a reduction in the ductility of the finished steel.

Ferrotitanium is added to the ladle for its effect on grain size and impact properties of certain steels. It is used especially in medium manganese steels to give them a fine grain over a broad range of heat-treating temperatures.

TABLE XX.—LOG OF ACID OPEN-HEARTH HEAT

Size of heat: 145,000 lb.

Composition desired: C, 0.25 to 0.30 per cent; Mn, 0.75 to 0.85 per cent; Si, 0.30 to 0.40 per cent.

11:00	2,000 lb. sand added to banks and bottom
11:10–11:40	First part of charge added
12:30–2:00	Second part of charge added
5:15	Melted
5:19	First test—C, 0.72 per cent, viscosity, $6\frac{1}{8}$ in.
5:25	300 lb. ore
5:54	Second test—C, 0.66 per cent; viscosity, $4\frac{7}{8}$ in.
6:15	1,000 lb. ore
6:41	Third test—C, 0.39 per cent; viscosity, $3\frac{1}{4}$ in.
6:54	Fourth test—C, 0.30 per cent; viscosity, $2\frac{3}{4}$ in.
7:09	Fifth test—C, 0.23 per cent; viscosity, 2 in.
7:12	500 lb. limestone
7:30	Sixth test—C, 0.18 per cent; viscosity, $2\frac{1}{2}$ in.
7:47	Seventh test—C, 0.17 per cent; viscosity, $2\frac{1}{2}$ in.
7:57	Eighth test—C, 0.15 per cent; viscosity, $2\frac{1}{4}$ in.
8:02	1,160 lb. spiegel
8:12	200 lb. burnt lime
8:25	Ninth test—C, 0.19; viscosity, $2\frac{3}{4}$ in.
8:30	1,750 lb. 80 per cent FeMn and 1,160 lb. 50 per cent FeSi
8:40	Tap

Ladle analysis: C, 0.28 per cent; Mn, 0.83 per cent; Si, 0.35 per cent.

Calcium silicon or calcium manganese silicon is added to the ladle as a final deoxidizer by some foundries. These deoxidizer alloys are used in connection with aluminum to produce a more desirable type of inclusion,

thereby promoting ductility. Additions will range from 2 to 5 lb. of the deoxidizer per ton of steel.

The tapping temperature of the metal will range from 2850 to 3000°F., depending on the size, number, and character of the castings to be poured. The temperature is judged by the aid of the film or set test (see Metal Tests, page 108). Each type of steel will form a film at various times according to analysis and temperature. These times are correlated and the goals are set in conformity with the number of castings and type of steel.

The log of an excellent acid open-hearth heat is given in Table XX.

Slag Control.—Very little has been published concerning systematic test-control methods on acid slags. However, quite a few operators have developed methods of their own and are accomplishing good results more easily than in years past when the eye of the melter was the only means of control. Three methods of control used are

1. Viscosity measurements by means of the viscosimeter as used in basic-steel shops or a modified form of it

2. Slag-cake tests examined for surface appearance, soundness of test, and color of fracture

3. Specific-gravity measurements

All of these tests have their particular advantages and are being developed in various shops. Of the three methods, the viscosity test is preferred. It is a very simple test to make and is accurate for the same slag over wide ranges of variables including speed of pouring, temperature of spoon and viscosimeter, and height of pour. The record on the heat card is tabulated in inches of flow.

Slag formation differs somewhat in the acid furnace from that of the basic furnace, since lime is charged at the beginning of the basic heat to form a slag, whereas in the acid furnace nothing except sand is added for the special purpose of producing a slag when the heat is melted. The result is that the acid furnace has much smaller volume of slag than the basic. The slag formed upon melting is made up almost entirely of the oxides of elements contained in the metallic charge—notably oxides of iron, manganese, and silicon, with the amount of silica augmented by erosion of the furnace banks and bottom. It is true that most operators will add sand to the bottom and banks prior to charging the heat. In most cases sufficient time is not allowed for this sand to fuse completely; hence a considerable portion presumably finds its way into the slag.

A typical acid open-hearth slag, when the bath is completely melted, will consist of approximately 50 per cent SiO_2 and the other 50 per cent FeO plus MnO. An actual analysis is as follows: SiO_2, 48.9; CaO, 1.5; Fe_2O_3, 1.3; FeO, 34.2; Al_2O_3, 1.8; MnO, 12.2 per cent. The high iron oxide content of the slag gives it a black appearance.

As the carbon content of the bath is lowered by the action of the iron oxide, the iron oxide in the slag is reduced and the slag color changes from black to brown to a light-green color, the tint of which is affected by the amount of manganese oxide present.

In some cases, to obtain a slag of proper fluidity, limestone or lime is added, and it may be desirable to hasten the conditioning of the slag by adding carbon in the form of coal to reduce its iron oxide content. A final slag may have a composition similar to the following: SiO_2, 58.2; CaO, 8.1; FeO, 19.2; Al_2O_3, 4.0; MnO, 10.2 per cent. If a high content of MnO is maintained in the slag, a composition similar to the following may be obtained; SiO_2, 54.1; CaO, 4.2; FeO, 13.3; MnO, 25.8; Al_2O_3, 3.3 per cent.

Slag Viscosity.—The control of acid slags has been primarily one of studying their viscosity or the manner in which slag strings out from a spoon test. Some acid open-hearth shops use the Herty viscosimeter throughout the heat. On other acid open-hearth floors the viscosimeter is not used since the slag, especially the finishing slag, is so viscous it will not run in the Herty viscosimeter. For these shops a modified viscosimeter has been constructed.[43] A practice that uses 5 to 8 per cent lime in the finishing slag is desirable if readings are to be recorded on the Herty viscosimeter. Fluidities of 2 to 3 in. can be obtained on finishing slags.

Probably the best method of controlling the viscosity of the slag is to use the proper amount of CaO or MnO and to regulate the temperature so that not too much silica is picked up from the furnace lining during the course of the heat. By regulating the silica pickup, the slag may be kept at the proper fluidity until the heat is ready for deoxidation. FeO and MnO are interchangeable as far as viscosity is concerned. It is possible, therefore, to deoxidize the slag without interfering with viscosity, provided manganese is used as a deoxidizer.

The viscosimeter can be used at the melt-down to estimate the time of ore addition. This is an important feature, because if ore additions are made to a slag that is too fluid, it is difficult to get the slag in proper shape thereafter.

It is reported that one foundry has found the viscosimeter to have a very important function in the working of the heat in the early stages. It reports that slags that run 4 in. or more indicate a heat that has not come up to temperature, in which case further additions of ore are not made until the slag reaches the proper temperature.

Most of the acid open-hearth shops that attempt slag control rate the viscosimeter as the important test and believe that it gives an indication of the acidity of the slag. The data shown in Fig. 15 indicate the relation between the acidity of the slag and its viscosity. A slag that runs 2 in. will in most cases have a silica content close to 60 per cent. The FeO

content of the slag will of course depend on the MnO and CaO contents of the slag.

Karmarov[14] has reported a study of the viscosity of acid open-hearth slags. He used the $\frac{1}{4}$ in. diameter bore mold recommended by Herty for basic slags. This work indicated the variation of viscosity with SiO_2 content of slags, and in general the work agrees fairly well with the curve of Fig. 15 and the work published by Juppenlatz and shown in Fig. 19.
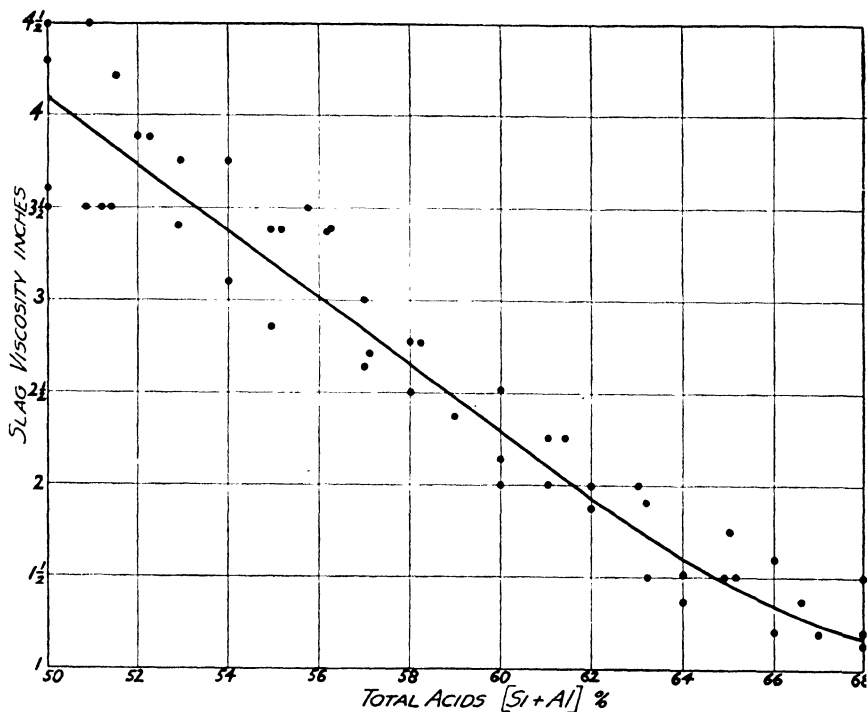
FIG 15.—The total acids (SiO_2 plus Al_2O_3) of an acid open-hearth slag as measured by slag viscosity. (*Briggs and Baldwin.*[2])

The relation between slag viscosity and the total FeO content of the slag when standard charges are used is shown in Fig. 16. Even though the charge and additions are standard, it will be noted from the scattering of the points that FeO content does not depend only upon the slag viscosity. There is an excellent trend indicated, however. It should be pointed out that this curve will apply only to the conditions of the foundry recording the data or possibly to those with closely similar practices.

An attempt is being made by one foundry to keep the viscosity of the finishing slag on steel casting heats above $2\frac{1}{4}$ in., which means that the FeO content will be at or above 25 per cent. This FeO content is rela-

tively high, since it is believed that high FeO is necessary to obtain and maintain a steel of good fluidity during the casting of the heat.
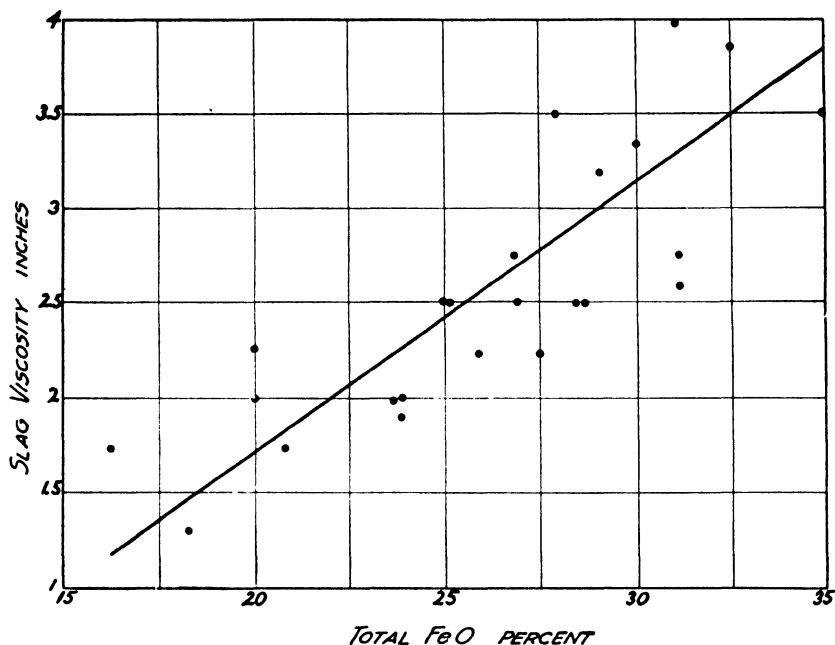
FIG. 16.—Relationship between the slag viscosity and the FeO content of acid open-hearth slag. (*Briggs and Baldwin.*[2])

Finishing slags for casting heats are sometimes kept under a 2-in. viscosity and over 1⅜ in. This means that the FeO content is below approximately 21 per cent and over 12 per cent. The low iron oxide slag

TABLE XXI.—SLAG-VISCOSITY CHANGES IN A 130,000-LB. ACID OPEN-HEARTH CASTING HEAT

Time	Additions, lb.	Carbon content, per cent	Slag viscosity, in.
5:05	Melted	0.76	3¾
5:12	700 ore		
5:42		0.53	4⅜
5:56		0.41	2⅜
6:03	400 limestone		
6:20		0.23	2¾
6:29		0.19	2½
6:33	1,040 spiegel		
6:37	180 lime		
6:44		0.20	2¼
6:49	1,040 FeSi		

permits a heat of good cleanliness with excellent manganese recovery. The FeO content of the slag is maintained over 12 per cent to prevent hydrogen absorption by the inactive bath. An example of the drop that may be obtained in the viscosity of the slag during an acid open-hearth heat of 130,000 lb. for steel castings is shown in Table XXI.

The color of the slag is not a good indication of its FeO content since high MnO contents mask the color effect. If low MnO contents are used, then it is quite possible that the slag color can be correlated to the FeO content.

An indication of surface appearance and fracture color of the slag cake as a function of viscosity and composition, as obtained from records on casting heats, is given in Table XXII. The degree of soundness of a slag cake is often studied by melters as a better indication of the FeO content than color.

TABLE XXII.—ACID OPEN-HEARTH SLAG-CAKE CHARACTERISTICS

Pancake			Viscosity, in.	Composition, per cent			
Top surface	Bottom surface	Fracture color		SiO$_2$	FeO	MnO	CaO
Rough...............	Rough	Black	7¼	....	....	....	....
Rough...............	Rough black	Black	5¾	45.3	37.9	14.7	0.45
Smooth dull..........	Smooth dark	Green	3¼	50.4	29.6	10.7	5.60
Smooth dull..........	Smooth dark	Green	2½	54.5	24.9	10.7	5.90
Smooth shiny.........	Smooth shiny	Green	2¼	55.6	21.6	12.9	7.30
Smooth shiny.........	Smooth shiny	Green	2	56.7	20.5	12.3	7.50

It is believed by some that the surface appearance and color of the slag test do not have much significance. However, in Fig. 17, four slag cakes are presented that were taken at different times in the heat. Viscosity tests were also taken at the same times. The log of Table XXIII

TABLE XXIII.—VISCOSITY CHANGES IN AN ACID OPEN-HEARTH HEAT
Charge: 139,000 lb.

Time	Activity	Slag cake	Viscosity, in.	Per cent		
				C	Mn	Si
2:11	Melt-down	No. 1	4¾	0.78	0.15	0.06
3:33	½ hr. after last ore addition	No. 2	3¾	0.28	0.07	0.05
3:51		No. 3	3	0.14	0.04	0.04
4:12	6 min. before deoxidation	No. 4	2¼	0.13	0.06	0.05
4:30	Tap (finishing analysis)			0.28	0.73	0.37

1   2

3   4
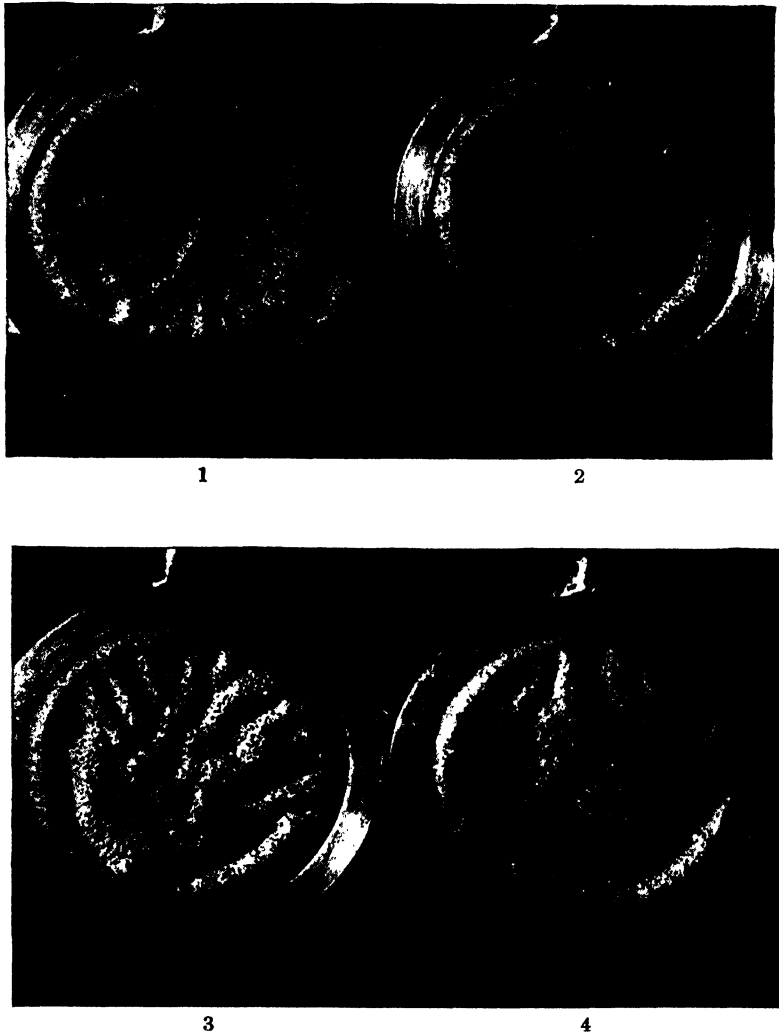
Fig. 17.—Acid slag-cake appearance. (*Briggs and Baldwin.*[2]) Viscosimeter tests were taken at the same time, and corresponding viscosities in inches and FeO contents of the slag are:

Slag cake	Viscosity, inches	Total per cent FeO
1	4.25	36.64
2	3.75	33.17
3	3.00	30.53
4	2.25	25.84

gives times and viscosities for the various slag-cake tests.   The results indicate that there are possibilities in such classifications.

It is the experience of acid open-hearth operators that slag analyses are not of much assistance if results cannot be reported within 25 min. However, it is true that an acid slag will come to practically a constant silica content for a given temperature at a given time.   Apparently, not too much information is available concerning the relation between the slag analysis and bath conditions.   At present it seems that the two can vary over rather wide limits without affecting the quality of the heat. However, when considering the limits set for heats made in the same shop, the relation between slag analysis and bath conditions varies over a comparatively narrow range.   The differences in furnaces, fuels, charges, products, etc., must all be considered when comparing results from shop to shop.

A study made of finishing slags for steel castings wherein a fluid steel is desired shows SiO_2 to be 53 to 56 per cent.   In pouring large castings where fluidity is not so important, the SiO_2 may range from 56 to 62 per cent with iron oxide content as low as 12 per cent.   It has been claimed that the higher the FeO content in the slag, the better the surface of the castings; consequently, it is the practice in two foundries to tap heats with as much as 30 per cent FeO.

It appears that for castings, acid open-hearth slags containing an FeO content from 15 to 25 per cent are quite satisfactory for the purpose intended.   Acid slags tend to reach an equilibrium rather quickly, and the quantity of silica and metal oxides present will depend somewhat on the temperature of the slag.   The equilibrium can be upset to a degree by the addition of iron ore; but within a short time equilibrium conditions will again be approached as the slag takes up silica from the banks.

Apparently the viscosity test is the best slag-control test for the acid open hearth.   It gives the operator an indication of the acid-element content of the slag, and it also gives a rough measure of the FeO content. Furthermore, it helps the melter to maintain a slag of the proper consistency, a procedure difficult to accomplish when relying entirely upon its appearance as observed by the unassisted eye.

Acid-furnace Life.—One of the most important items in the life of the acid furnace is the great rapidity with which the uptakes or topmost checkerwork become choked.   Serious choking usually begins at the top of the air regenerators and largely determines the length of the campaign.   When the furnace schedule is interrupted at intervals for cleaning the checkerwork, the ensuing cooling off reduces the effective life of the furnace brickwork.   Choking is due to an accumulation that changes in nature from a hard sponge adhering to the topmost bricks to a fine dust at lower temperature levels in the chamber.   These accumula-

tions consist of almost pure ferric oxide. Some furnace operators try to clean the checkers every 100 to 125 heats by compressed air and long blowpipe. The accumulations are allowed to collect on the bottom of the chamber until the end of the campaign.

The slag pockets can be cleaned more easily and in some cases are cleaned six to eight times during the campaign. The weights of slag removed will total 180,000 to 250,000 lb. per campaign for a 60-ton furnace. It has been reported that the more agitation that is set up in the outgoing gases in the slag pockets, the less material goes over into the checkers. Agitation can be increased by using a straight-faced bridge wall.

ACID-ELECTRIC PROCESS

The number of electric-arc furnaces in the steel casting industry is considerably greater than that of other melting units, and most of these furnaces are operated by the acid practice. In September of 1944 there were 186 foundries in this country operating acid-electric furnaces. The total number of acid-electric furnaces active and being installed in these foundries was 337. A production of approximately 1,300,000 net tons of steel castings was made by the acid-electric method in 1943. Upon the basis of only one heat per day—a conservative calculation—the molten-steel capacity was 2,100,000 tons per annum. The 3-ton rated-capacity furnaces are the most popular, with 82 in operation.

The furnaces are but seldom charged at the rated-capacity figures. Standard charges are usually from 1.3 to 2 times the rated capacity of the furnace. Consideration should be given to overcharging the furnace, for by so doing the quality of the steel may be affected. The ratio of the bath area to the weight of the charge is a more important figure in determining furnace overcharging than any value based on the rated capacity of the furnace.

The acid-electric process for steel castings is a fast-oxidizing practice that does not permit the elimination of phosphorus or sulphur from the charge. The elements carbon, manganese, and silicon can be reduced to a low figure in a relatively short time. Since phosphorus and sulphur cannot be even partially eliminated by the process, it is necessary that care be exercised in the selection of the charge, especially when producing steel to phosphorus- and sulphur-specification limits. This requirement has caused but little difficulty in the past, since there has been available plenty of scrap with phosphorus and sulphur contents below 0.05 per cent each. The scrap situation during the period of the Second World War was not so favorable, and all acid-electric foundries experienced difficulty in obtaining scrap to the phosphorus and sulphur specifications desired. In fact, a few foundries found it necessary, because of the scrap situation, to change to the basic practice.

The reasons for the preference of the acid-electric process by steel foundrymen are, according to Finster,[15] as follows:

1. The flexibility it offers in producing small quantities of steel to meet specifications of various chemical and mechanical-property requirements.

2. Lower refractory cost, due to lower initial cost of the lining material and to its longer life.

3. Lower power consumption when compared with a two-slag basic practice.

4. Greater tonnage output per unit time for a given charge.

5. Slag of higher viscosity than basic slag, which is an important consideration, since in many foundries making small castings, the molds are poured from top-pour hand shanks and a viscous slag is more easily controlled.

6. Greater fluidity of the metal, permitting the casting of thinner sections with greater assurance of having all corners and details clearly reproduced.

The actual manipulations of the process have not been standardized, and hence it is quite possible that there are as many variations as there are foundries producing acid-electric steel. These variations are in general the result of adapting the acid practice to the individual requirements of the furnace. Minor modifications of the practice will not be discussed, and only the major variations will be reported.

There are four major variations of the acid-electric process as used for the manufacture of steel castings: (1) the partial-oxidation, (2) the complete-oxidation and the single-slag, (3) complete oxidation with silicon reduction, and (4) the double-slag practice.

The Furnace.—There are about six manufacturers producing electric furnaces in this country. The majority of the arc furnaces are of the three-phase, a-c type, equipped with automatic electrode control, and can be tilted. Transformer ratings generally vary from 100 to 4,000 kva., depending on the size of the furnace. The primary-line voltages vary from 4,000 to 26,400 volts, most foundries having about 13,000 volts. The total amperes per phase used varies from about 2,000 to 10,000. The more modern types of furnaces have at least three and sometimes four combinations of voltage. These secondary no-load voltages between phases are most frequently around 200 for the high tap, 160 for the intermediate, and 120 for the low tap. The older furnaces have only two secondary voltages, such as 150 and 90 volts. The secondary no-load voltages lower than these figures are also being used.

The guide to the electrical operation of an arc furnace is the circuit characteristics shown in Fig. 18.[16] The maximum power (kilowatts) in the circuit occurs at 0.707 power factor. The maximum power in the furnace, *i.e.*, the maximum rate of heat development, occurs at some

higher power factor, a value that is dependent on the resistance and reaction of the conductors of the circuit. There is an individual set of values for optimum current and maximum power, efficiency, and power factor for each applied voltage. These values can usually be determined by trial toward the end of a heat when the circuit is stable and balanced. The slope of the power-current characteristic on each side of the optimum current value is small and the term "optimum current" means, in practice, a range of current values below and above the actual value.
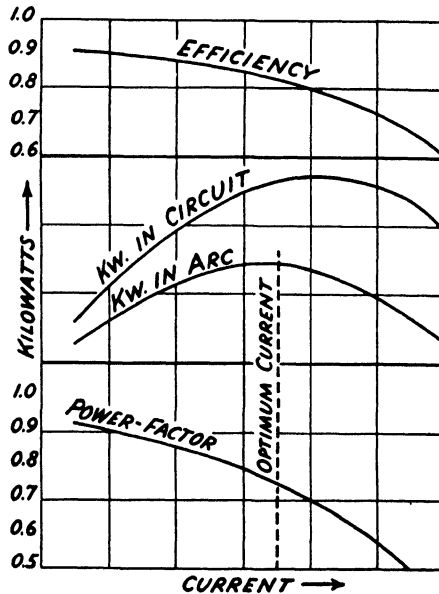
FIG. 18.—Circuit characteristics of an arc furnace. (*Reprinted by permission from "Industrial Electric Heating" by N. R. Stansel, published by John Wiley & Sons, Inc.*)

Some value of current below the optimum current value may give the desired rate of heating. A current value higher than the optimum value is undesirable. The value of current selected for use must, of course, be within the rating of the electric apparatus regardless of what may be the value of the optimum current.

There are two types of electrodes being used: (1) graphite and (2) carbon. It is the opinion among foundrymen that the larger diameter carbon electrode results in a smoother operation than can be obtained by the smaller graphite electrode of equal electrical efficiency. The graphite electrodes are the more costly of the two types, but this extra cost is offset somewhat by the greater ease in handling the smaller graphite electrodes, as well as by the lower freight costs; furthermore, the storage space can be smaller.

Some foundries have observed that they can detect no difference in the operations between the two types of electrode. For example, 8-in. carbon and 6-in. graphite electrodes have been used interchangeably in sets on the same 1½-ton furnace. The most common electrode sizes for steel foundry use are the 4-, 6-, and 8-in. graphite electrodes and the 8-, 10-, 12-, and 14-in. carbon electrodes. There is apparently an advantage in the use of the graphite electrode when producing stainless steels or steel of low-carbon content, since there is less chance of pickup of carbon from the electrodes by the bath.

Monthly records on electrode consumption for various foundries throughout the country show rather wide limits of 9 to 23 lb. per net ton charged for carbon electrodes, and 5 to 15 lb. per net ton charged for graphite electrodes. An individual foundry will not show these wide variations if values are reviewed over a period of months. An average consumption for 6 months will show, for example, 10.1 lb. per net ton charged, while the upper and lower values during this period will be 9.5 and 10.6 lb. per net ton. The consumption is considerably influenced by the rate of production, type of steel made, size of the furnace, and such special melting practices as increasing the carbon content of the bath by dipping the electrodes.

Furnace Linings and Refractories. *Roof.*—In most cases the roof of the acid furnace is constructed by the use of standard- or special-shaped silica brick. During periods of greatly curtailed operations and intermittent melting schedules, fire-clay-brick roofs have been used advantageously. With intermittent melting schedules, considerable deterioration is due to spalling that results from heating and cooling the furnace, and the clay brick stand up as well as the silica brick under such conditions. For best performance of clay roofs, it is advisable to raise the roof several inches to allow for the inferior refractory characteristics of clay. During periods of continuous melting, the silica-brick roofs give the greater life.

Silica brick should be purchased to specifications as follows:

Chemical analysis:

	Per Cent
SiO_2	94.5
Al_2O_3	2.0
CaO	2.5

Pyrometric cone equivalent: At least Seger cone 32 (3110°F.)
Total porosity: 25 per cent
Specific gravity: No brick to be over 2.38, and the average of the consignment to be less than 2.36

It has been observed that an acid-furnace roof that takes an even glaze during the early stages of its life is more likely to have a longer life.

Observation should be made of the brick-surface conditions, since under precisely similar heating conditions some brick show no sign of glazing, some glaze evenly, and others show a rough surface because the grog (refractory particles) remained solid while the matrix melted. The matrix contains the bond and also various easily ground impurities, and these affect the glazing characteristics. Spalling action on the roof is most severe around the holes through which the electrodes pass.

Side Walls.—The side walls of the acid-electric furnace are lined either with silica brick or rammed ganister (quartzite). Most foundries use silica-brick linings; however, there is a noticeable tendency for foundries to change to the newer type of rammed linings.

The silica-brick lining is built up from the bottom of the furnace shell in one or two courses, up to the removable top. Very few foundries employ the practice of filling in a layer of loose insulation material of about 2 in. in thickness, between the steel furnace shell and the silica-brick lining.

The rammed linings usually consist of graded ganister as a base. Their consistency is the same as that used for the rammed bottom. Forms are prepared for ramming and the material is tamped in, to a thickness of from 8 to 15 in. By using the rammed material, a new bottom and side-wall lining can be built up in about one-fourth the time required to put in a brick lining. In a few instances, it is reported that the bottom is ganister and side walls are ganister to just above the slag line. Above this, the side walls are silica brick. In most cases, the door linings are rammed with a good grade of plastic fire clay.

Bottom.—Most foundries prefer a brick construction next to the shell, since this gives a substantial foundation for the brick side-wall lining. This construction constitutes a lifetime bottom and is not removed when new rammed-bottom linings, which are laid upon it, are introduced. These bottom brick are usually silica brick, although in some instances the heavy-duty fire-clay brick are used. If the side walls are of rammed construction, the bottom brick are often left out so that the entire lining is rammed. Not all the bottom material is removed when it is necessary to put in a new rammed bottom as a base.

The bottom material is carried up the side walls to above the slag line. A few foundries have a preference for sintered silica sand as a bottom material. This bottom is built up in much the same way that the sintered acid open-hearth bottom is constructed. The furnace is heated with a coke fire; then the bottom is built up by small additions of sand, and the electrodes arc on coke or pieces of electrodes on the bottom. This is a slow procedure since the electrode contact material must be raked out of the way periodically so that additional sand may be placed on the bottom. Exceptional care must be taken in order that a

well-constructed bottom be prepared. A bottom that is easily sintered into place is one that contains 5 per cent acid slag mixed with the sand, with an addition of 2 per cent ferrosilicon added to the last 2 to 3 in. of bottom laid.

A rammed lining makes a far better bottom than a poorly sintered one; hence the majority of foundries use the rammed bottom. Moreover, the time required for constructing the rammed bottom is but a fraction of that necessary for preparing a sintered bottom. The base material of the rammed linings is either crushed ganister or silica sand. The majority of the foundries use the crushed ganister.

If sand is used, it is usually bonded by mulling it with fire clay and a small percentage of water to which has been added glutrine or molasses. Ganister is generally purchased to a sizing specification. One hundred per cent of the material should pass a 4-mesh screen. Some foundries prefer the material to be finer by having it all pass through an 8-mesh screen. The ganister is bonded in a number of ways. In some cases it is bonded by a small amount of water to which has been added an organic material such as molasses. In other cases an addition of 6 to 8 per cent molasses (no water) is mulled with the ganister, or fire clay up to 20 per cent is mulled with ganister and a small percentage of water is used. A few foundries prefer to bond the ganister by the addition of a small amount of pitch.

Patching material for holes that develop is usually the same as that from which the bottom is prepared. The ganister should have the following chemical limits: SiO_2, 95 per cent minimum; Al_2O_3, 1.50 per cent maximum; Fe_2O_3, 2.50 per cent maximum; CaO, 0.50 per cent maximum.

Lining Life.—Lining life is quite variable and depends on a number of factors, such as the care taken in lining installation and maintenance, the size of the charge used (the degree of overloading), and the operating conditions, since intermittent operations shorten refractory life. The customary roof life is about 300 heats. A study of 24 acid-electric furnaces of varying melting capacities gave the following information regarding lining life:

	Heats per roof	Tons melted per roof	Heats per lining	Tons melted per lining
Average of 24 furnaces..............	288	830	264	755
Average of 4 furnaces melting 5 to 6 tons.	320	808	305	869
Average of 13 furnaces melting 3 to 5 tons	283	973	252	943
Average of 7 furnaces melting 1 to 3 tons.	282	516	260	503

This set of data shows that the roof life exceeds that of the side-wall lining. The bottoms are repaired after each heat, if necessary, and rebuilt when new linings are put in. The reason for the shorter lining life as compared with the roof life is probably the use of rammed lining by some foundries. One foundry reports 400 heats per silica-brick lining and only 225 heats for a rammed ganister lining. Other foundries are able to obtain the same life for the rammed lining as for the roof.

The effect of overloading upon the lining life is illustrated by the report of one foundry that in charging 5 tons per heat the refractory life was 900 tons per roof and lining, whereas, in charging 4 tons per heat, the life of the roof was increased to 1,000 tons per roof and the lining to 1,200 tons.

The Charge.—The acid-electric-arc furnace charge usually consists of foundry scrap (gates, risers, and reject castings) and purchased scrap. The purchased scrap will consist of heavy and light plate and plate shearings, forgings, stampings, rails, springs, punchings, flashings, turnings, and bale scrap. The scrap should contain low phosphorus and sulphur contents, and every endeavor is made to procure basic open-hearth or basic electric scrap. Some steel castings, such as coupler parts (basic open-hearth steel), are used when available.

The foundry scrap, consisting of gates, risers, and defective castings, constitutes from 15 to 50 per cent with the majority of the foundries charging 30 to 40 per cent of foundry scrap. A study of 30 acid-electric foundries showed that they averaged a 38 per cent foundry-scrap charge. Percentages greater than 50 per cent foundry returns can be used. Charges of 100 per cent foundry scrap can be melted successfully, provided due consideration is given to the higher silicon content of the charge in working the heat. Tests at one foundry showed no low mechanical properties after increasing the foundry scrap in the charge to 100 per cent. Some foundry operators are of the opinion that they cannot use their own foundry returns in quantities of much more than 30 per cent of the charge, or they will run into low mechanical properties. Such conditions will not be encountered if the complete-oxidation practice is used.

All acid-electric foundries give particular attention to the character of the scrap purchased and to its composition. Phosphorus and sulphur contents must be below 0.05 per cent, since some specification maximums are set at 0.05 per cent and oxidation losses on melting tend to concentrate the phosphorus and sulphur slightly. The welding grades of steel have restrictions on residuals, which constitute a problem requiring further scrap analysis and selection care. Limits have been placed on molybdenum, chromium, copper, and nickel, since each of these has an effect upon the hardenability of the heat-affected base metal during welding.

The foundryman must also give considerable thought and care to the composition of his own foundry returns and scrap castings.   Most acid-electric foundries produce from 5 to 15 types of standard alloy steel.   The gates and risers of these various grades should be kept separate or marked in such a manner that they can be added to charges being prepared for similar alloy heats.

A few foundries add 2 to 3 per cent pig iron to the charge.   This increases the carbon content of the charge so that a more vigorous boil will be obtained after melting.   Such additions are usually made when the carbon content of the charge averages only 0.25 to 0.30 per cent.   A low-phosphorus pig iron is obtained for this purpose.

In charging the furnace, care must be taken that the charge is fairly compact so that an arc may be maintained between the electrodes and the charge.   The heavy scrap should be placed on the furnace bottom beneath the electrodes; and the lighter scrap such as turnings, flashings, and light sheared plate should be around the edges against the banks and the lining.   Heavy scrap placed on top of the charge may cause electrode breakage.

The kilowatt-hour, electrode consumption, and the melting time per ton are influenced by the condition of the charge.   Loosely stacked charges are responsible for increasing all three items.

It is a decided advantage of the swinging-roof furnace that it allows the scrap to be arranged in drop-bottom charging buckets in the same manner that it should lie in the furnace.   Top charging can be accomplished quickly, the entire operation taking not more than 1 or 2 min.   The entire charge is usually added at one time.   In some cases, when the scrap is quite bulky, this may mean that the furnace will be three-fourths full of scrap metal.

Steelmaking Methods.—There is no uniformity of opinion as to the best method to be adopted in producing acid-electric steel for castings. In 1929, George Batty[17] stated that

. . . some operators hold that the best steel is produced by adjusting the components of the charge to produce a certain carbon content without having to induce a heavy boil, or even an appreciable boil, to bring the carbon down to the desired figure. Other steelmakers insist that the better method is to ensure that every heat of plain carbon steel for castings be boiled, which involves the elimination of a considerable portion of the carbon of the charge.

A study (in 1941) of the melting operations of 30 foundries showed that the method of oxidation of the heat was not so controversial a subject, since 27 of the 30 foundries were found to maintain a long, vigorous boil.   The method of using the vigorous boil is known as the "complete-oxidation" method; the process used by the minority of the acid-electric

foundries is referred to as the "partial-oxidation" method. The advantages and disadvantages of the two methods are as follows:

Partial Oxidation

    Advantages:
        1. Rapid production of steel
        2. Good refractory life due to low FeO content of the slag
        3. Low power input
        4. Low electrode consumption
        5. No recarburizer required
    Disadvantages:
        1. Insufficient removal of gases
        2. Occasional lack of fluidity
        3. Occasional poor mechanical properties

Complete Oxidation

    Advantages:
        1. Low included-gas content
        2. Good fluidity
        3. Good mechanical properties
        4. By elimination of nearly all the silicon and most of the manganese, a uniform base metal is obtained from which final additions can be calculated closely.
    Disadvantages:
        1. Slower production of steel
        2. Short refractory life because of a corrosive slag high in FeO content
        3. Higher production cost

    Partial-oxidation Process.—In the partial-oxidation process the desired charge is one that will have a carbon content of about 0.05 to 0.10 per cent above the finishing carbon. If the average carbon content is higher than this, then a small amount of iron ore—a quantity of not more than 0.5 per cent by weight of the charge—will be added to the charge, or some very rusty scrap will be used. No slag-making materials are added to the charge, and a fast-melting practice is used.

    The normal amount of rust present on the scrap, together with the oxide resulting from oxidation due to the heat of the arc, is sufficient to produce a light boiling action during melting. By the time two-thirds of the scrap has melted, a light, gentle boil may be noticed on the bath. When melting is complete, the carbon content of the bath will be about 0.25 to 0.35 per cent, and the silicon content about 0.15 to 0.22 per cent.

The actual value will depend on the degree of oxidation, such as the use of iron ore or very rusty scrap in the charge.

By the time the carbon and manganese preliminaries are reported to the melter, the carbon has dropped a point or two; and by the time the desired temperature is reached, the carbon is low enough, so that by the addition of the ferroalloys the carbon pickup from these alloys will bring the carbon content into the finishing range desired.

Additions are seldom made to the slag. The slag will be about 3 to 5 per cent by weight of the bath and will consist of sand adhering to the foundry scrap used in the charge, together with SiO_2 that has eroded from the furnace banks, side walls, and bottoms.

Approximately 5 min. after the deoxidizers are added to the furnace in the form of ferromanganese and ferrosilicon, the furnace is tapped. It has been observed that foundries that use this type of practice only occasionally have mechanical test-property minimum specifications to meet. They use a good grade of scrap, and their primary object is to get as high a rate of production of molten steel from the furnace capacity as possible. There is a considerable quantity of low-priced steel castings that do not require any acceptance tests other than superficial surface inspection, and foundries that cater to this type of work normally select the partial-oxidation process, since it is the cheapest.

Complete-oxidation Process. Melting and Oxidation Period.—The great majority of all steel foundries using the acid-electric practice have found it necessary to employ the complete-oxidation process. Nearly all of the cast steel produced in these foundries must pass minimum mechanical specification requirements, and many of the castings are subjected to radiographic and magnetic inspection. The process consists of melting the charge at high voltages. During the melting operation and the following boiling period the oxidizing reaction $C + FeO \rightarrow Fe + CO$ occurs, in which the iron oxide is supplied by the rust on the charge and by iron ore or mill scale added to the bath.

Slag-forming materials, other than the sand sticking to gates and risers, are not added to the charge. Some operators add a small amount of iron ore or mill scale to the charge or make this addition when the charge is about one-quarter to one-half melted. The majority of operators wait until the charge is completely melted, or very nearly melted, before adding iron ore or mill scale. The latter method is preferred as this will result in a bath with a more vigorous boiling action.

The melting-down carbon will fall anywhere from 0.18 to 0.40 per cent. The majority of those using the process will melt down with a carbon content of around 0.20 per cent. The manganese content at melt-down will be 0.25 to 0.35 per cent, and the silicon content will be 0.13 to 0.20 per cent for the standard practice, which calls for the addition of iron

ore after all the charge is melted.   The manganese and silicon contents will be lower than the above values if iron ore is added with the charge or added during the melt-down period, the values obtained depending on the degree of oxidation.   Another disadvantage to the practice of adding iron ore to the charge and during melt-down is that an estimate of the amount needed cannot be obtained; indeed, this change can only be obtained when the charge is melted and the bath analysis is determined.   The use of this practice may result in overoxidation because of excessive iron oxide additions.

Most foundries prefer to use iron ore (hematite) as the iron oxide addition.   A few foundries prefer to use mill scale in preference to iron ore, since it can be spread across the bath more quickly, causing a general boiling action.   Also, mill scale is easily obtainable at a low cost.   The users of mill scale, however, warn that care must be exercised in its procurement, since it may contain higher sulphur contents because of high sulphur-bearing fuel oils used in the heating of billets for forging.   The use of high-sulphur-content mill scale will increase the sulphur content of acid-electric steel.

The amount of iron ore used ranges from 0.75 to 4 per cent of the weight of the charge.   The average is about 1.25 to 1.75 per cent.   The amount of iron ore added will vary according to the carbon content of the charge and the state of oxidation of the purchased scrap.   In some instances the stock piles of scrap are such that uniform charges may be prepared over a considerable period of time.   Under this condition the iron-ore addition can be made standard.   It is possible that a complete-oxidation practice be followed without the addition of any iron ore or mill scale owing to the heavily oxidized character of the purchased scrap and a high proportion (approximately 75 per cent or more) of its use in the charge.

Chromium ore and manganese ore are used at times as oxidation agents.   The chromium ore is used, of course, only in the manufacture of chromium-bearing cast steels.   The addition of iron ore to the melted charge will produce a vigorous boiling action and the carbon, silicon, and manganese contents of the bath will be materially reduced.   It is the practice of some foundries to oxidize the carbon content down to approximately 0.10 per cent.   The actual carbon content at the end of the boiling period is not so important as the maintaining of a vigorous boil for 15 to 25 min.

The carbon drop or the rate of carbon drop will indicate whether or not this requirement has been carried out.   For example, one foundry may melt down at 0.30 per cent carbon and at the end of the boiling period show a carbon content of 0.15 per cent in the bath.   Another foundry may melt down at 0.20 and after boiling show a bath carbon

content of 0.10 per cent.   Both heats have maintained an adequate boil for the complete-oxidation process, for as the carbon content of the bath decreases, the length of boil is increased in order to oxidize additional carbon.   With this point in mind, it can be seen that the carbon content of the bath at the end of the boil may be anywhere from 0.08 to 0.18 per cent carbon.   One group of steel melters prefer a carbon content of 0.10 per cent; an equally large group of steel melters aim at 0.15 to 0.18 per cent as an end point.

It should also be pointed out that a practice of complete oxidation can be employed even though the carbon content of the bath at the end of the boil is greater than 0.18 per cent as an end point.   A few foundries maintain a vigorous boil and finish the boil with carbon contents from 0.20 to 0.25 per cent.   The silicon, however, is below 0.10 per cent.   The complete-oxidation practice depends on a vigorous boil for approximately 15 to 25 min.   If it is desired to finish the boil with a higher carbon content, then the melt-down carbon must be equally higher.

The silicon content at the end of the boil will vary from about 0.02 to 0.10 per cent.   The lower values usually are recorded by those who oxidize the carbon to about 0.10 per cent.   The manganese content will be from 0.07 to 0.15 per cent, depending on the manganese content of the charge at melt-down.   The importance of a low residual silicon and manganese in the bath prior to deoxidation has been pointed out by Batty[17] and Sims[18] and has become a recognized practice by many acid-electric operators.   The best acid-electric practice calls for maximum residual of silicon at 0.08 per cent and manganese at 0.15 per cent.

A high residual manganese, for example, above 0.15 per cent, greatly increases the danger of skin porosity.   This is probably not a direct effect of manganese but because a high residual manganese reduces the oxygen content of the steel and causes a more gentle boil.   Gases are not effectively eliminated from the bath upon the employment of light boils.   A high residual silicon or maintaining for a long period those conditions that cause the reduction of silicon from the slag is more effective than high residual manganese in producing gassy steel for similar reasons.

During the latter part of the oxidation period, additions of slag-forming material may be made to the bath.   Sand may be added to increase the slag bulk and thereby to reduce the FeO content.   Also, lime may be added to release some of the FeO combined with the silica in order to continue a gentle boiling action.   However, most slag-forming materials are added during the refining period.

It has been reported by Porter[27] that in order to obtain maximum recovery of oxidizable alloys present in the charge, two points must be observed, as follows:

1. The percentage of the alloying elements introduced through the scrap charged must be kept low.

2. The working of the heat must be done in such a manner that the possibility of overoxidizing the heat is avoided and the carbon content after proper oxidation is as high as possible, consistent with the finishing analysis of the steel.

If these points are used as a basis of operation, Porter[27] reveals that when manganese in the charge is under 0.90 per cent, with sufficient carbon present so that after proper oxidation the carbon content is between 0.30 and 0.40 per cent, the residual manganese is in the vicinity of 40 per cent of the manganese originally charged. If, however, the oxidation is such that the carbon content at the end of the boil is between 0.10 and 0.20 per cent, the residual manganese averages approximately 25 per cent of the manganese in the charge. Also, if the manganese added is from 1.25 to 1.50 per cent of the charge, the residual manganese found after oxidation is about 30 per cent of the manganese charged in the 0.30 to 0.40 carbon range and 18 per cent in the 0.10 to 0.20 per cent carbon range. Silicon in properly oxidized heats is always below 0.10 per cent. Chromium present in amounts up to 2.00 per cent in the charge is recovered about 70 to 75 per cent. If larger amounts of chromium are present, the recovery drops off. Hence larger percentages should not be in the charge if a saving in valuable alloys is to be effected.

Refining Period.—Toward the end of the oxidizing period the boiling action will slow down, unless too much iron ore was added, causing an overoxidized slag. With the residuals low, the power is cut down, allowing the boil to subside. Recarburization is begun and slag additions are made. Some foundries prefer to add all the recarburizer at one time, although it appears more acceptable to add about half of the recarburizer and then, as the carbon boil slows down, to add the remainder.

There are a number of recarburizers used by acid-electric operators. These are: low silicon, low phosphorus, pig iron, or wash metal; spiegeleisen; broken electrode pieces; petroleum coke; flake graphite; breeze coal; and high-carbon steel. Or carburizing is effected by dipping the electrodes. Pig iron is most commonly used, although the practice of dipping the electrodes is favored by some operators. It is believed that the best results are obtained by the use of pig iron or spiegeleisen, since a more uniform and a better carbon boil is developed over the bath. The carbon in the pig iron combines with the FeO in the metal in the following manner:

$$C + FeO \rightarrow CO + Fe$$
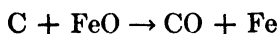
The CO rising out of the bath produces the boiling action, known as the "carbon boil." The carbon boil is very effective, since it reduces

the FeO content in the metal toward the equilibrium value, which is another way of saying that the possible content of nonmetallic inclusion is reduced. Moreover, the carbon boil is an additional security against the inclusion of hydrogen and nitrogen, since these gases are able to leave the bath only through the formation of CO bubbles. The carbon boil developed by dipping the electrodes is more concentrated in the area near the electrodes and as a rule does not extend over the entire bath.

There is a wide variation in the amount of slag-forming materials added at this time. In a number of foundries no slag-making materials are used. It is believed that the quantity of slag should not exceed much more than 6 per cent by weight of the bath. Sand additions may vary from 4 lb. per ton (0.2 per cent) to 40 lb. per ton of charge (2.0 per cent). If limestone is added, the amount varies from 1.5 lb. per ton to 5 lb. per ton. If lime is used in place of limestone, the quantity will be about one-half the amount of limestone normally used. It is best to add sufficient lime or limestone to produce a CaO content of the slag of from 6 to 8 per cent. Perhaps about only two-thirds of the foundries that follow the complete-oxidation practice make lime or limestone additions. A survey of 24 foundries showed that 17 used CaO additions while 7 did not. The CaO in the slag reduces the FeO content of the slag. In some cases fluorspar is added to increase the fluidity of the slag.

The slag is allowed to come to the proper consistency, and at the same time the temperature of the bath is increased. A few furnace operators assist the reduction of FeO in the slag by the addition of a small amount (0.03 to 0.05 per cent by weight of the charge) of pulverized ferromanganese, scattered across the top of the slag. Powdered coke is but seldom used in the deoxidation of the slag because of its effect in lowering the FeO content of the slag to such a point that silicon is reduced from the slag into the metal. If powdered coke is added, it is usually added late in the refining period and carefully scattered over the slag.

The carbon content is checked to see whether it is near the range desired, and further additions of pig iron are made if necessary. After the proper slag color and viscosity are obtained and the temperature reached, the alloy additions are made, such as chromium, vanadium, or molybdenum, and then the deoxidizers are added.

Deoxidation.—The primary object of deoxidation is the prevention of reaction between dissolved oxygen and carbon to produce gaseous products. Any element that is capable of reducing FeO in molten steel may be considered a deoxidizer. Deoxidizers vary in their ability to reduce FeO, although no deoxidizer is able to reduce all of the FeO present in molten steel. This matter will be discussed in detail in Chap. III.

Most foundries deoxidize the heat by first adding ferrosilicon, followed 1 to 3 min. later by an addition of ferromanganese. A study of 42 steel

foundries using the acid-electric practice showed that only two add the ferromanganese first.   Several foundries add silicomanganese in preference to ferrosilicon and then correct to the proper manganese and silicon requirements by small additions of ferromanganese and ferrosilicon to the furnace or the stream during tapping.

The total time taken for the addition of the deoxidizer is from 4 to 10 min.   The deoxidizer is added just prior to tapping, and then a few tests such as a set test, porosity test, or a fluidity test may be taken, but this is done as quickly as possible so as not to delay the heat.

The ladle is brought into place at the time the first deoxidizer is added, so that it will not be necessary to hold up the heat by waiting for the ladle.

Tapping.—A few foundries follow the practice of adding a part or all of their ferromanganese to the stream in the furnace spout prior to its entering the ladle.   This is done to obtain a higher recovery of manganese.   The slag is held back during pouring, allowing the manganese to be taken up by the metal without reacting with the FeO of the slag. Care should be taken during the tapping operation to prevent the slag from entering the ladle with the metal.   Tapholes or skimming spouts are constructed so that when the furnace is tilted over quickly, the metal and not the slag will enter the ladle first.

The steel is tapped at temperatures from 2900 to 3200°F.   In nearly all cases the tapping temperature is in excess of 3000°F.   This is because of the large number of castings that must be poured in one heat; the high temperatures are necessary so that thin sections may be successfully cast.

Special Deoxidizers.[1]—Nearly all acid-electric steel melters add special deoxidizers to the ladle.   These are necessary because acid-electric steel is primarily used for the pouring of small castings, the molds for which are made of green sand.   The pouring of small molds requires the use of hot metal.   Although the metal is dead-killed[2] on leaving the furnace, the exposure of molten metal during tapping into a bull ladle and then into hand shanks and during the pouring into the mold is considerable.   Each transfer is accompanied by an increase in surface exposure; and the green sand mold itself appears to oxidize the metal.   To correct these conditions and thereby to prevent porosity, certain special deoxidizers are added to the ladle, to the hand shanks, and to the stream as it enters the mold.   Aluminum is the most widely used of all the special deoxidizers.   A review of the melting practice of 26 acid-electric foundries showed that 25 used aluminum.

[1] In the steel casting industry, deoxidizers other than ferrosilicon and ferromanganese are termed "special" deoxidizers.

[2] The reaction of oxygen with carbon is prevented because of an adequate presence of silicon.

More than half the foundries using the aluminum added it to the ladle. The others added it to hand shanks or to both ladle and hand shanks. Some reported adding it to the molds also. It is believed that the best place for addition is to the ladle, with the aluminum in ingot or strip form, tied to a rod and plunged to the bottom of the ladle after all the metal has been tapped into the ladle. However, all foundries do not add it in this manner. In some cases a small (⅜ in.) length of aluminum rod will be added to each hand-shank ladle prior to filling with metal from the bull ladle. Others prefer the use of shot or rod aluminum added to the metal stream as it leaves the furnace.

Aluminum is added in amounts of from 4 oz. to 3.5 lb. per ton. An average of 26 foundries showed the use of 1.5 lb. per ton. During the past few years foundries have made a considerable change in the quantity of aluminum added, as a result of the findings of Sims and Dahle;[6] most melters are now requiring the addition of more than 1.5 lb. of aluminum per ton.

Other special deoxidizers such as calcium silicon, calcium manganese silicon, calcium silicide, and ferrotitanium may be added with the aluminum. Usually they are added just prior to the aluminum addition. Probably only about one-quarter of those who use aluminum also use additional special deoxidizers. These deoxidizers are added for special protection and for the development of certain types of inclusion (see Chap. III). The quantities of these deoxidizers used vary considerably according to individual needs. These are about as follows:

	Lb. per Ton
Calcium manganese silicon	1.5–3
Calcium silicon	2–3
Calcium silicide	2–3
High-carbon ferrotitanium	2–5

Operating Time.—The oxidation and the refining time vary from 30 to 55 min. in most of the acid-electric foundries. An average for 20 foundries showed 41 min. The oxidation period runs from 15 to 25 min., with the remainder the refining period, which also includes the time necessary for deoxidation. The average heat time per ton of melt varies from about 20 to 70 min., the average from 24 foundries being 34 min. per ton of melt. The average kilowatt-hours per ton of melt also varies between limits of about 450 to 700 kwh., with the average for the 24 foundries of 570 kwh. per ton. An average of 3.70 man-hours per ton of melt was recorded for the entire melting department, including supervision and ladle maintenance.

One foundry claims 17 min. heat time per ton of melt;[29] 2,500 to 3,000 kv. input of 235 volts and high reactance are employed. A 9,800-lb. charge is melted in 47 min. The boil is 12 min., refining is completed

in 9 min., and the heat is tapped and recharged in 10 min.—or a complete cycle from tap to tap in 80 min., with a production of 3.7 tons per hour. An average of 3.4 tons per hour or better is desired by the foundry.

Heat Log.—A typical heat log for an acid-electric furnace operation based on the complete-oxidation, single-slag process is shown in Table XXIV. The log shows four tests for separate chemical determinations. This is greater than the average number of tests usually made although those foundries that are equipped with an electrical method of carbon

TABLE XXIV.—FURNACE LOG OF A HEAT OF CARBON STEEL MADE BY THE COMPLETE-OXIDATION PRACTICE

Size of heat: 5,000 lb.
Composition desired:
    C, 0.25 to 0.30 per cent; Mn, 0.65 to 0.75 per cent; Si, 0.30 to 0.40 per cent

Charge:

	Lb.	Per cent
Foundry scrap	2,000	40
Plate	1,250	25
Punchings	1,000	20
Bale scrap	750	15

Time	Activity
8:00	Completed charging (hot furnace)
9:08	Melted down
9:08	First test—C, 0.22 per cent; Mn, 0.26 per cent; Si, 0.16 per cent
9:10	Iron ore, 100 lb. (2 per cent)
9:20	Second test—C, 0.16 per cent
9:30	Third test—C, 0.12 per cent; Mn, 0.07 per cent; Si, 0.04 per cent
9:30	Low-silicon pig, 100 lb.
9:33	Lime, 10 lb.; sand, 50 lb.
9:38	Low-silicon pig, 100 lb.
9:42	Fourth test—C, 0.21 per cent
9:47	Slag test—blue-green
9:50	Ferrosilicon, 40 lb.
9:52	Ferromanganese, 50 lb.
9:55	Tapped

Aluminum, 50 lb. (2 lb. per ton) added to the ladle.
Ladle Analysis: C, 0.27 per cent; Mn, 0.69 per cent; Si, 0.35 per cent.

determination may take from three to four tests, since carbon can be determined fairly accurately in 2 to 3 min. The residual-manganese content is reported in a number of foundries, and the silicon content is reported only occasionally—whenever the character of the scrap changes. Samples for the determination of nickel, chromium, and molybdenum are taken at melt-down of the charge, since their determination takes from 20 to 25 min.

Slag Control.—The acid-electric process is generally a one-slag process. Some operators let the furnace make its own slag, stating that by so doing the slag volume can be held to a minimum and that a higher manganese yield can be obtained from its final additions. Other operators add sand to protect the furnace refractories from attack by the basic oxides formed, while still others decrease the FeO content of the slag by lime or limestone additions.

Melt-down slags contain from 35 to 55 per cent FeO and from 35 to 55 per cent silica. Finishing slags[15] in practices where no lime is added contain FeO, from 15 to 20 per cent; MnO, 15 to 25 per cent; and SiO_2, 57 to 62 per cent. Where lime is used, the CaO content is generally held between 4 and 10 per cent, and the FeO content of this type of finishing slag is lower—about 12 to 17 per cent. If the double-slag practice is used, the FeO content of the finishing slag carrying CaO is usually between 7 and 13 per cent. An example of typical slag analysis for a carbon-steel heat is shown in Tables XXV and XXVI.

TABLE XXV.—SLAG ANALYSES FOR A TYPICAL CARBON-STEEL HEAT. COMPLETE-OXIDATION PRACTICE. VIGOROUS BOIL

Time, min.	Test	Metal, per cent			Slag, per cent				Viscosity, in.
		C	Mn	Si	SiO_2	FeO	MnO	CaO	
0	Approximate charge	0.25							
70	At melt-down	0.20	0.30	0.07	50.3	38.6	9.5	None	6
90	End of boil	0.10	0.09	0.02	43.3	42.3	12.8	None	7.5
100	After adding SiO_2 and CaO.								
	After recarburization	0.20	0.10	0.05	51.2	27.7	12.0	6.1	4
111	Before deoxidizers	0.20	0.11	0.05	60.2	16.6	12.4	7.8	1.75
115	At tapping	0.25	0.68	0.41	59.1	14.3	16.3	7.1	2.0

Analyses not made for Al_2O_3 and MgO. Iron oxide calculated as FeO.

TABLE XXVI.—AN 8,000-LB. CARBON-STEEL HEAT. SHORT, LIGHT BOIL Juppenlatz[20]

Time, min.	Test	Metal, per cent			Slag, per cent				Viscosity, in.
		C	Mn	Si	SiO_2	FeO	MnO	CaO	
0	Approximate charge	0.27	0.53	0.19					
90	At melt-down	0.23	0.17	0.05	42	37	12	7	9
103	Before boil	0.22	0.15	0.04	48	30	13	7	5
108	After boil	0.17	0.15	0.02	52	27	12	7.5	3
111	Before deoxidizers	0.20	0.16	0.04	57	22	12	7.5	2.5
115	Before tapping	0.26	0.73	0.40	57	18	15	8.5	2
118	Ladle	0.26	0.66	0.32	56	16	18	8.5	2.25

A tabulation of 10 foundries that conducted slag studies showed that the range of analyses of the finishing slag was as follows:

	Per Cent
SiO₂	56–62
MnO	{10–17, 22–27}
CaO	3– 8
FeO	10–15

Four foundries preferred the 22 to 27 per cent MnO range.

There are a number of ways in which slag conditions are observed in the acid-electric foundry. The usual manner is to observe the color of a sample of cold slag. Both the surface and fracture colors are studied. Also, the structure of the fracture surface is observed. The manner in which the slag strings out from a rod or spoon is an indication of the slag viscosity. Some operators make actual viscosity tests and others density tests. These tests are used to assist the operators in estimating the FeO and SiO₂ of the slag.

Occasionally, chemical analyses are made for FeO, SiO₂, and CaO. These slag analyses require considerable time and are not of much aid to the melter during the making of the heat. Other tests that can be correlated with slag analyses are of more practical value.

The viscosity of acid slags is related to its chemical composition; slag-viscosity tests may be made and from them the composition inferred, the results of which are immediately available to the melter. A number of acid-electric steel melters are now consistently using the Herty viscosimeter (¼-in.-diameter flow channel) on all heats. The slag sample should be taken with a clean spoon at a point in the furnace about halfway between the electrodes and the banks. The sample should be quickly withdrawn and inverted into the funnel of the viscosimeter. The quickness of this operation is important, since the temperature of the slag affects its viscosity.

The viscosity test gives an indication of the SiO₂ content of the slag and a general indication of the FeO content. The relationship of the silica content to viscosity is shown in Fig. 19. The curve is taken from the studies of Juppenlatz,[20] and the points indicated are average values reported by Finster for 65 heats.

The FeO content as plotted against viscosity[20] shows a considerable scattering of points (Fig. 20). However, there is a definite trend toward higher fluidity at the higher FeO values. The total iron oxide, FeO plus Fe₂O₃, is expressed as FeO.

The MnO and CaO contents of the slag do not control its viscosity; Juppenlatz[20] shows that only flat curves could be constructed if CaO and MnO contents are plotted against viscosity. The addition of lime in the

ranges normally used in acid-electric practice does not appreciably affect the silica-viscosity curve, and for practical purposes this curve may apply to lime-free slags as well as to lime-bearing slags.
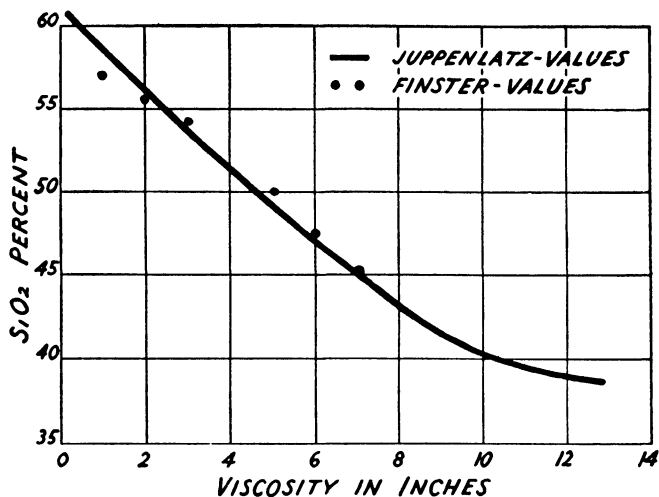
Fig. 19.—Relationship of silica content to the viscosity of acid-electric slags. (*Juppenlatz*[20].)
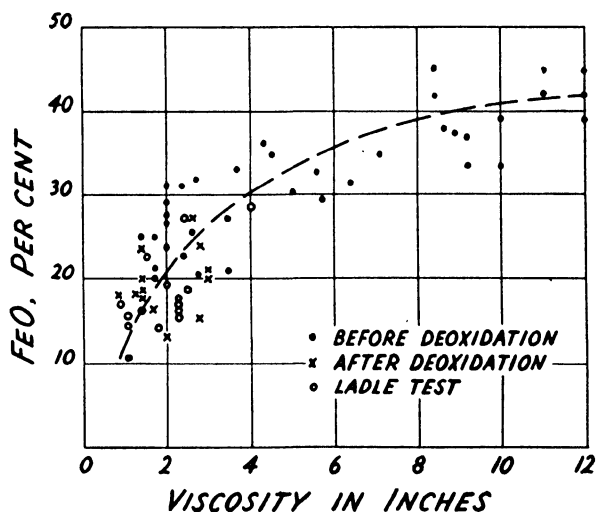
Fig. 20.—Relationship of iron oxide content to the viscosity of acid-electric slags. (*Juppenlatz*[20].)

Viscosity values reported from foundries show that slag conditions may be considerably varied and still produce good heats of acid-electric steel. One reported method is to melt down under a very fluid slag of about 10-in. viscosity, which corresponds to approximately 40 per cent

FeO.   Prior to deoxidation the slag would run about 1 to 2 in., a slag viscosity corresponding to 10 to 20 per cent FeO.   Another practice is to melt down under a heavy slag with a viscosity ranging from 2 to 4 in. During the oxidation period the slag is thinned with ore and lime additions so that a viscosity of from 5 to 8 in. is obtained.   The heat is finished when the slag reaches a viscosity of 1 to 2 in.

It is generally true that the manganese and silicon losses are rather great at the time of deoxidation if the finishing slag has a viscosity of 4 in. or greater.   A slag with a low viscosity of 1 in. at the time of deoxidation will allow high recoveries of manganese and silicon.   Special care must be given to the handling of low FeO slags, since silicon can be reduced readily from these slags, leading to a silicon pickup by the metal and the possibility of producing steel of poor fluidity and low mechanical properties.

Slag-cake Tests.—It is the practice of some melters to judge the iron oxide content of the slag by observing the appearance and color of a sample of the solidified slag.

Slags with over 40 per cent FeO have a crystalline fracture of a dull gun-metal color with a surface skin of the same color.   As the FeO content decreases to 30 per cent, the fracture becomes amorphous with a gray to dull-black color and a dull-black surface.   Slags with 30 to 20 per cent FeO have a dark-green fracture and a dull-black surface. Between 20 and 15 per cent FeO, with normal CaO and MnO contents, the dark green either separates as streaks in a lighter green matrix or becomes a solid jade-green color.   The surface color changes from a dull black at 20 per cent FeO to a shiny dark brown at about 15 per cent FeO. With normal CaO and MnO contents, all slags with less than 15 per cent FeO are a uniform blue-green with the skin changing to a light shiny brown as the FeO content approaches 10 per cent.   Under 10 per cent FeO, the brown skin becomes thinner until the green interior shows through.   At 5 per cent FeO, very little brown skin can be seen and the slag is a uniform light blue-green (light jade) throughout.

The slag colors change considerably, depending on the MnO and CaO content of the slag.   Increasing the MnO content will make the slag appear as though there were a much lower FeO content than is actually present.   The effect of CaO is more marked and therefore more confusing.   Table XXVII shows the effect of various SiO_2/CaO ratios on the appearance of the slag with quite wide variations in FeO, as prepared by Caine.[21]   The slag samples were taken from all stages of the heat.   The effect of the FeO is to darken the green color in the opaque jade-green slags until it is a very dark gray-green at 24.6 per cent FeO. Contents of FeO up to 20 per cent have no visible effect on the translucent glassy slags.

TABLE XXVII.—SLAG-CAKE COLOR WITH VARYING SiO_2/CaO RATIOS
Caine[21]

SiO₂/CaO ratio in slag	FeO content	Slag fracture color
2.70– 3.37	7.1 –11.0	Opaque jade or apple green
3.74	9.2	Mixture. Mostly opaque jade green but with some translucent dark green showing
4.15– 4.38	9.0 –11.8	Mixture. Mostly dark translucent glassy green with some opaque jade green showing
4.55– 4.63	9.0 –19.3	Dark translucent glassy green
4.80– 5.24	10.0– 12.0	Mixture. Mostly dark translucent glassy green with some opaque jade green
5.24– 5.61	10.84–13.14	Mixture. Mostly opaque jade green with some dark translucent glassy green
5.61–12.80	11.0 –24.6	Opaque jade green

It is possible from the slag-cake appearance to obtain qualitative information of the FeO contents of the slag. However, it is neither accurate nor sufficiently foolproof to enable the melter to hold the FeO content of the finishing slag within a narrow range such as one of 3 per cent, as required by a few foundries. Such range requirements dictate the use of the double-slag process.

Double-slag Process.—The double-slag process is practiced regularly by some acid-electric foundries that are particularly concerned with keeping the FeO content of the finishing slag to about 10 per cent. It is believed that since refining periods are so short, and since slag-cake and viscosity tests are not sufficiently accurate for the estimation and control of FeO content, a positive method of FeO control should be adopted. This can be accomplished by removing the oxidizing slag and adding slag-making materials in proper proportions.

The double-slag process is identical with the complete-oxidation process to the point of recarburization. At this time, the oxidizing slag is removed, and sand and lime are charged to the furnace in definite proportion. The quantity of slag-making material charged is equivalent to about 2 to 5 per cent by weight of the charge, and the lime content of the slag from 6 to 10 per cent. The slag materials are melted as quickly as possible, and then after the usual refining period the heat is deoxidized in a manner similar to that described in the complete-oxidation process. The FeO content of the finishing slag can be held at a low value within a narrow range by this process. It permits a uniformly high recovery of manganese and silicon at the time of deoxidation.

Typical slag and metal analyses during the course of a double-slag heat are given by Caine[21] in Table XXVIII. By the use of the double-slag process, it is possible to maintain a constant volume of slag, heat

after heat, thereby favoring standardization of heats.   The double-slag practice is used occasionally by most foundries when excessive portions of the bottom come up or because of poor slags resulting from poor scrap. The double-slag process is generally more time-consuming, on an average of from 10 to 15 min. longer than the single-slag process.

TABLE XXVIII.—TYPICAL DOUBLE-SLAG PRACTICE.   SLAG AND METAL ANALYSES
Caine[21]

Test	Metal analysis, per cent				Slag analysis, per cent						Temperature, °F.
	C	Mn	Si	FeO	SiO₂	FeO	Al₂O₃	CaO	MnO	MgO	
Charge.........	0.20	0.50	0.20								
Test before heat is completely melted. Oxidizing slag....	0.16	0.27	0.04	0.207	47.0	45.0	3.5	None	4.2	0.5	2820
End of ore boil..	0.06	0.05	0.06	0.191	48.6	42.5	3.2	None	5.7	None	2850
Second slag. Before dipping electrodes.....	0.07	0.07	0.04	0.175	44.2	31.8	6.0	7.6	4.8	3.9	2900
After dipping electrodes.....	0.12	0.08	0.06	0.150	46.7	26.9	6.1	7.7	4.6	4.2	2900
Second slag. After FeSi addition........	0.12	0.10	0.35	0.140	54.6	19.2	6.5	7.4	3.5	3.8	2980
Second slag. After FeMn addition......	0.19	0.50	0.40	0.112	67.6	15.1	6.7	8.5	4.6	3.0	3000
Ladle slag.......	0.21	0.55	0.41	0.055	61.6	10.5	8.4	8.8	5.8	3.1	2890

Carbon is added with Mn and Si for higher C heats

A time increase is not always necessary, for, as pointed out by Caine, (the foremost advocate of the double-slag processes), studies made on kilowatt-hour consumption and time of heats made under his direction showed that results compare favorably with averages given previously on the complete-oxidation process.   Records for one furnace show 571 kwhr. per ton and 33½ min. per ton, which agree exactly with the averages for the single-slag method (570 kwhr. per ton and 34 min. per ton). In general, Caine believes that the double-slag process is more costly and time-consuming than the single-slag process but that the increase in cost is very small and that the over-all cost, including losses from rejections, is lower.

Silicon-reduction Method.—A reduction of silicon from the slag to the metal can be accomplished in the acid open-hearth practice as well as by the acid-electric practice. However, the higher temperatures employed in the acid-electric melting, the possibility of using more viscous slags at these high temperatures, and the high concentration of heat input under the electrodes all favor reduction of silicon from the slag in the acid-electric practice.

European acid practice is usually one that specifies silicon reduction. It is thought that a small amount of silicon reduction is necessary to the production of "good" acid steel. It is the belief in this country that silicon reduction during the production of an acid heat gives rise to more trouble than can be offset by the economic considerations arising from dispensing with the ferrosilicon addition. Reference has been made repeatedly, in American publications on steel castings, to silicon reduction and the influence of "overreduced" steel[1] on fluidity, porosity, and low mechanical properties. The method used to correct this condition is to add iron ore to the bath or the slag.

The overreduced condition arising from silicon reduction is an unintentional part of the steelmaking process and is brought about by maintaining high temperatures over unduly long periods with low iron oxide–content slags. Even though the problems arising from overreduced steel are appreciated, it is standard practice in some foundries for economical reasons to reduce all the silicon from the slag and not to add any silicon in the form of ferrosilicon. The most essential point in the silicon-reduction process is to be certain that an active boil is maintained during the oxidizing portion of the heat, so that the silicon-residual content is below 0.10 per cent.

The refining of the heat proceeds in a manner similar to that in the total-oxidation process, except that the temperature of the bath is increased materially toward the end of this period. A bath temperature of 3100 to 3200°F. is most effective for rapid silicon reduction. The more rapid the silicon reduction from the slag to the metal, the higher will be the mechanical properties. In some cases the entire silicon reduction takes place in the last 10 min. of the heat. The silicon content of the metal is raised during this period from about 0.10 per cent to 0.30 to 0.55 per cent.

The silicon reduction takes place during a slight boil, the reaction being

$$2C + SiO_2 \rightarrow Si + 2CO$$
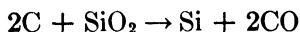
[1] The term "overreduced steel" is employed to express a condition of subnormal fluidity found occasionally in the running of a molten steel. This so-called "overreduced" condition may be remedied by a suitable addition of iron oxide to the molten steel. Hence, a steel in unsatisfactory condition can be improved by oxidation, and the unsatisfactory state is more or less logically designated as "overreduced."

This reaction takes place only at high temperatures and when the FeO of the slag is low. Most melters believe they can tell when the silicon reduction is taking place, since copious white fumes issue from the furnace around the electrodes. Blosjo[22] counsels against depending on the white fumes as a telltale of silicon reduction, since he has experienced silicon reduction without the presence of the fumes.

TABLE XXIX.—FURNACE LOG OF A HEAT OF CARBON STEEL MADE BY THE SILICON-REDUCTION METHOD

Blosjo[22]

Size of heat: 4,000 lb.
Composition desired:
  Carbon, 0.25 to 0.30 per cent; manganese, 0.70 to 0.80 per cent
  Silicon, 0.35 to 0.45 per cent

Charge:

	Lb.	Per cent
Foundry scrap	1,800	45
Plate	1,300	32
Steel castings	200	5
Springs	700	18
	4,000	

Elapsed Time, Min.	Activity
0	Completed charging (hot furnace)
20	30 lb. old slag added (0.75 per cent)
52	Iron ore—20 lb. (0.50 per cent)
65	Melted down—first test—C, 0.32; Mn, 0.18; Si, 0.10 per cent
66	Slag test No. 1—FeO, 28.28; MnO, 17.57 per cent
74	Second test—C, 0.29; Mn, 0.15; Si, 0.08 per cent
	Slag test No. 2—FeO, 24.25; MnO, 18.30 per cent
80	Third test—C, 0.22; Mn, 0.16; Si, 0.09 per cent
	Slag test No. 3—FeO, 17.00; MnO, 20.08 per cent
86	Fourth test—C, 0.19; Mn, 0.18; Si, 0.11 per cent
	Slag test No. 4—FeO, 14.97; MnO, 19.30 per cent
96	Fifth test—C, 0.14; Mn, 0.23; Si, 0.23 per cent
	Slag test No. 5—FeO, 14.24; MnO, 18.00 per cent
101	Pig iron—126 lb. added
102	FeMn, 0.80 per cent—36.5 lb. added
105	Heat tapped

Ladle addition:
  Aluminum 2 lb. per ton
Ladle analysis:
  C, 0.26; Mn, 0.79; Si, 0.42; P, 0.037; S, 0.039 per cent
Ladle slag analysis:
  FeO, 9.46; MnO, 19.35 per cent

A log of a heat prepared by the silicon-reduction method is tabulated by Blosjo[22] in Table XXIX and is shown in Fig. 21. In explaining the heat, Blosjo points out that the voltage plays a very important part; a rate of 325 kwhr. per ton per hour was used during the last 30 min. of the heat. The change in the chemical composition of the bath may be observed in Fig. 21. It will be noted from this chart that the major part of the silicon is reduced in the last 12 min.

The old acid slag that was added early in the heat acts as a cover during the melting-down process, and a high MnO content is necessary to reduce 0.30 to 0.50 per cent silicon. The material contains about 10 per cent FeO and 20 per cent MnO and therefore makes a rapid, low-
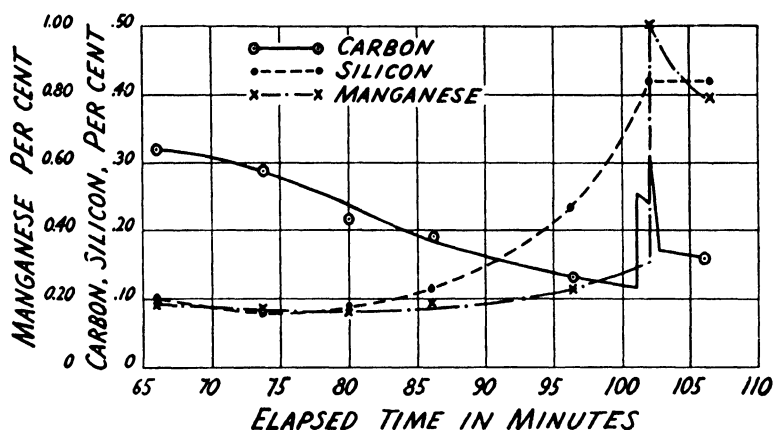
Fig. 21.—Change in chemical composition of the bath during the making of a carbon-steel heat by the silicon-reduction method. (*Blosjo*[22].)

melting slag cover. The MnO is kept at a high value in order to prevent too much oxidation of manganese during the melt-down and the boil. With this practice the manganese residual is seldom below 0.15 per cent. A high-MnO-content slag is influential in effecting a high-manganese recovery. The manganese recovery in the heat of Table XXIX is 77 per cent.

The manner in which manganese enters into the reduction reaction is not known. Apparently if the MnO content of the slag is about 5 to 8 per cent, a reduction of about 0.20 per cent silicon at 3200°F. is the limit that can be obtained. On 1.5 per cent manganese heats at 3200°F., it is possible to reduce 0.60 per cent silicon. The MnO content of the slag in this case is approximately 15 to 20 per cent.

The silicon-reduction practice requires considerable skill and unless an operator has at his disposal a furnace equipped with a high input, so that a very high bath temperature can be obtained in a relatively short interval, it is better that the method not be attempted.

Metal Tests.—After the final deoxidizers are added to the bath and just prior to the tapping of the heat, several metal tests are made by furnace operators. The most common test employed is the spoon or set test. This test gives an indication of the temperature of the steel and its fluid life. The test is made by slagging a small spoon and taking a sample of the bath about halfway between the electrodes and the banks. As the spoon is withdrawn from the furnace, a stop watch is started. The slag is removed from the top of the metal and a record is made of the time in seconds required for the metal to film over and to solidify. A correlation of these periods, with optical temperature reading at tapping, will give an indication of the temperature. The manner and the time in which the film forms give an indication of the fluid life of the metal.

A few foundries use a fluidity mold at the furnace. About 8 lb. of metal are taken from the furnace in a large spoon and dumped into a spiral mold as quickly as possible. The mold is then shaken out and the length of the spiral in inches indicates the fluidity of the metal.

A test for porosity is also taken at the furnace. A test bar of from $\frac{1}{2}$ in. in diameter and about 6 in. in length is cast vertically in a green sand mold with a sample of steel from the furnace. If the test solidifies with a definite shrink, the bar is broken and the fracture examined for pinhole porosity. If there is a tendency for the metal to rise in the mold, further additions of ferrosilicon and sometimes ferromanganese are made. If the metal continues to rise, the heat should be reboiled, the slag removed and a new slag put on, and treatment given according to the double-slag process.

These tests require only 1 or 2 min. after being set up. A cooperative furnace crew could make all three tests and record the results in the space of 2 or 3 min.

ACID CONVERTER PROCESS

The converter production of steel for steel castings in the United States is on an upward trend. Whether this is due only to the extra production that the industry was called upon to supply during the Second World War or to a revival of the use of converter steel for castings is

TABLE XXX.—CONVERTER PRODUCTION

Year	Gross Tons
1915	92,476
1918	160,844
1920	104,980
1925	53,834
1930	14,871
1933	3,012
1940	25,000*

* Estimated.

problematical, but it is believed to be the result of both. The change in production of converter steel is shown in Table XXX. The turning point came with the low production in 1933. It is believed that the high point of yearly production rate during the war was in excess of 50,000 gross tons.

In the years just prior to 1933, twelve steel foundries were using the converter process, employing 17 converters. In January of 1944, thirty-seven foundries in the United States were operating 66 converters. The Whiting Corporation, manufacturers of side-blown converters, reported that 21 converters were sold during the first 6 months of 1942.

The 2-ton converter is the most popular size installed in foundries. The capacity and number of converters in the industry, in production and on order in January, 1944, are shown in Table XXXI.

TABLE XXXI.—NUMBER AND CAPACITY OF CONVERTERS

Capacity, Tons	Number
1	16
1.5	5
2	25
2.5	3
3	12
4–6	5

Several factors are responsible for the revived use of the converter other than the need for steelmaking units to assist in production for the Second World War. These factors are

1. The melting of an all-steel charge by the modern hot-blast cupola, thereby permitting the selection of a charge low in phosphorus and sulphur.

2. Improved methods of desulphurization of cupola metal that permit production of converter steel with sulphur contents between 0.025 and 0.045 per cent.

3. The phosphorus content of the blown metal can be reduced by a ladle treatment with a basic oxidizing slag provided the steel is further treated in an electric furnace.

4. Adaptation of the photoelectric cell control for accurate determination of the end point of the converter blow, which makes the training of operators easier and substitutes instrument control for that of the human eye.

5. Development of a triplexing method, using cupola-converter-electric-furnace combination, which permits continuous pouring operations.

6. A better understanding by the consumer and the general public of the mechanical properties, including low-temperature impact properties, of steel produced by the converter process.[30]

In addition to the advantages resulting from the developments that have been made in the technique of cupola-converter operation and in

the understanding of the properties of converter steel, the side-blow converter has other advantages. The installation cost of the side-blow converter is approximately one-fourth that of any other steelmaking unit of equal capacity. There is an insignificant demand charge for electric current for converter operation, whereas the high demand charges are a very serious factor with electric-furnace operation even though production in the foundry is small or intermittent.

One of the major features of interest in the converter is its flexibility in the production of carbon and low-alloy steels. It can be put into action on short notice, and small lots of various types of steel can be made at any time since orders of 1 or 2 tons would constitute an entire heat.

In common with other methods of steelmaking, certain limitations are to be found in the side-blow converter process. One limitation is the rather high melting losses, both in the cupola (3 to 6 per cent) and in the converter (10 to 20 per cent). Another limitation is the necessity of using metal of low phosphorus content in the cupola, since there is no phosphorus reduction in furnaces lined with acid (siliceous) linings. This is also true in acid-electric and acid open-hearth steelmaking processes. It is possible to obtain a reduction in phosphorus if the Yocom process is used or if certain phases are carried out in a triplex process in conjunction with converter practice. The Yocom process can be used for castings only if the steel is given a subsequent deoxidation treatment in the electric furnace, since the Yocom process leaves it in an oxidized condition.

In present-day manufacture of steel castings by the converter process the large bottom-blown vessel, known specifically as the "Bessemer converter," is not used. In the earlier days of the industry, the Bessemer converter[31] was used and foundrymen attempted to make use of small units. They were hampered by the difficulty of producing steel in small heats that was hot and fluid enough to run thin-section castings. In the efforts that were made to overcome this difficulty, changes were made in the converter construction to permit the warming up of cold heats by blowing at or over the surface of the bath by tipping the vessel forward. The extra metal temperature was attained by the combustion of iron and the burning of CO to CO_2 within the vessel. The Walrand-Legenisel process for attaining higher temperature metal was not used to any extent in this country, although it was fairly successful in France and England in the nineties. This process consisted of adding ferro-silicon after the drop of the flame and then reblowing for a definite period of time, thus producing higher temperature metal.

Converter Construction.—The side-blow converter, introduced into this country by Tropenas around 1898 and hence often referred to as the Tropenas converter, has the blast introduced through tuyères in one side of the vessel above or at the surface of the bath.

The converter consists of a cylindrical vessel made in two pieces (Fig. 22), both constructed of heavy steel plate with the shell and bottom section fitted with a short funnel-shaped top piece clamped to the shell. The shell is removable to facilitate lining. The vessel is supported by two trunnions mounted on standards so that it can be rotated to any position desired for charging and tapping of the metal. The air required for the operation is supplied by a positive pressure blower through a hollow trunnion into the tuyère box mounted on the shell of the vessel. The tuyères, five to seven in number, are 15 to 17 in. above the bottom lining. The tuyères are from 1 to $1\frac{1}{4}$ in. in diameter.
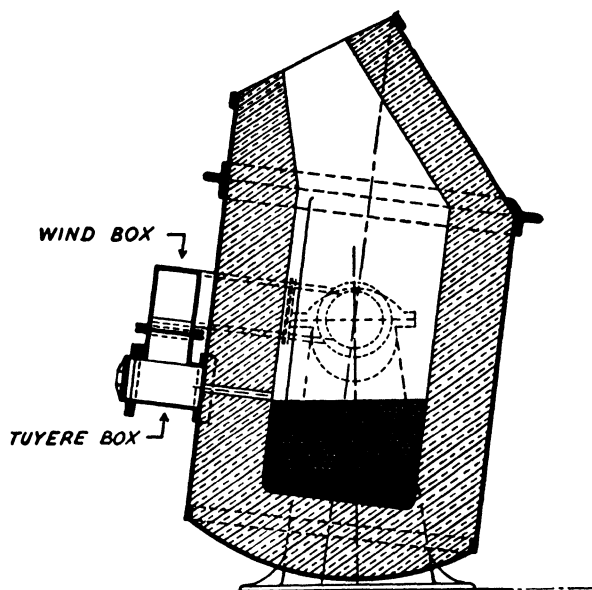
Fig. 22.—General construction of the side-blow converter.

Converter capacities and the corresponding blower capacity and other operation details are given in Table XXXII.

Each converter is provided with an exhaust hood to remove smoke, gases, and sparks produced during the blowing operation.

Lining.—The converter lining may be either of silica brick or a rammed-in ganister monolithic lining. In the United States, when the rammed lining is used, the converter usually is lined with about 9 in. of silica brick with a 9-in. rammed ganister lining laid inside, covering the brick.

The silica brick used for lining may be either standard shapes (9-in. straights and arches interspersed to form the desired circle) or shapes made up especially for lining converters. The special converter brick are usually made in two sizes, so that an inner and an outer circle of brick

can be removed and replaced when repairs are necessary. The tuyère
section is made in one complete shape, in two half-sections, or with one
brick for each tuyère opening. In using silica brick the customary
allowances for expansion must be made. Care must be exercised in the
preliminary drying of the lining, or the brick will spall and crack, par-
ticularly the larger tuyère shapes. Before starting operations the lining
should be inspected and cracks patched to prevent runouts of metal
during the blow.

TABLE XXXII.—CONVERTER CAPACITIES[32]

Capacity per blow, tons	Approximate capacity per hour, tons	Blower capacity, cu. ft. per min.	Blower motor, hp.	Air pressure, lb.	Matching cupola tons per hour, tons	Size of tuyère openings, sq. in.
1	2– 3	1,750	40	5–4	4	6.4
2	4– 6	3,250	75	5	7	12.5
3	6– 9	5,000	125	5	10	17.2
4	8–10	6,000	175	5	12	20.7
5	10–12	7,000	200	5	14	23.5
6	12–15	8,000	250	5	17	29.7

The ganister that is used for linings is obtained from quartzite, a
siliceous rock crushed and screened to a suitable fineness. The fineness
normally used is 100 per cent through 4-mesh (0.187 in.), with not much
more than 25 per cent through the 25-mesh screen as fines. The ganister
is mixed with 10 per cent high-quality fire clay and moistened with
sufficient sodium silicate solution, or an organic solution such as molasses
water, to give a good ramming bond. For best results the mixture should
be mulled in a mill and stored for 24 hr. before using, without allowing
it to dry out.

One foundry has set up a specification for lining material as follows:

Green shear strength.............................. 10.5 p.s.i.
Permeability.. 85–90
Moisture... 5.5 per cent

The lining material is rammed to the proper depth at the bottom of the
vessel. A substantial wooden form, conforming to the cross section of
the desired holding capacity of the vessel (about 18 in. in height and fitted
with a lifting ring), is placed on the bottom; the material for the side wall
is rammed around it. This form can be raised from time to time until
the lining is completed. The tuyères are formed by using pieces of pipe
of the proper dimensions, fastened in a holding device, which in turn is

clamped in the wind box. The thickness of the lining at the tuyère side may be greater and in some cases is 11 or 12 in. The tuyères must be exactly perpendicular to the axis of the vessel; an error of a degree or so will prolong the blow 5 to 8 min. and will yield cold steel. Upon completion of the lining, the pipe-holding device and the pipes are removed. Also, an excellent bottom may be obtained by lining with silica brick with the $4\frac{1}{2}$-in. side of the brick standing vertical.

The nosepiece of the converter is lined in the same manner, with another wooden form made to follow the desired contour. The two pieces are assembled and clamped together, the inner joint having more ganister rammed in to make a smooth-finished job and to prevent metal from leaking through the vessel. The cross ramming should extend vertically about 12 in. on each side of the joint.

The lining, whether brick or ganister, must be thoroughly dried, either by a coke fire or by a gas burner. A low fire is maintained for 10 to 15 hr. Some operators are of the opinion that a coke fire does not properly heat the section below the tuyères. These operators employ the use of a specially constructed water-cooled gas burner, which is placed in the nose of the converter with the flame directed toward the bottom. About 2 hr. before the first blow the lining should be heated as hot as possible, preferably by introducing oil or gas burners through the backing plate of the wind box and extending through the tuyères. Successful blows are very dependent upon having the lining of the converter brought to almost incandescent heat. With continuous operation the lining will lose very little of its heat. However, if it is necessary to delay operations between blows for some time, then the lining should be again heated before charging the converter with metal. Optical temperature measurements on the incandescent lining before the first blow are generally in the range of 2250 to 2300°F.

If oil is not burned to heat the converter, additional coke is added to the coke fire used for drying the vessel, and air is blown through the tuyères at $1\frac{1}{2}$ p.s.i., until the lining is near incandescent heat. The vessel is then turned down and the coke dropped.

Each blow erodes some of the lining of the vessel, necessitating daily patching, while extensive lining repairs are made each week end. The slag and eroded sections are chipped out sufficiently to expose the brick lining or undamaged rammed material, which permits the insertion of templates and ramming of 3 to 4 in. of new lining for the full length of the vessel. In continuous practice, an additional shell is advantageous, which, having been previously lined and dried out, can be used to replace the one in service. In one converter installation operating continuously for 16 hr., three shells are in use, one mounted on the trunnions in service, one being dried out ready for service, and one being repaired. Another

foundry advises that they require four shells for continuous converter operation.

The average life of a lining is from 30 to 60 blows. Lining life is shortened by (1) hotter blows, (2) longer blows, and (3) more manganese in the pig iron (because the resultant slag is more corrosive). Free iron oxide attacks silica rapidly; consequently, the attack around the tuyères is quite rapid.

Melting.—The size of the cupola used will depend somewhat on the converter size and schedule of operation. The cupola is charged with foundry returns of steel gates and risers, purchased steel scrap, low-phosphorus pig iron, ferrosilicon, limestone, and coke.

The charge may consist of from 70 to 85 per cent steel scrap and 30 to 15 per cent pig iron, the silicon content being adjusted by the use of silicon briquettes or 50 per cent ferrosilicon. An all-steel charge is often used. The make-up of this charge usually consists of 50 per cent foundry returns and 50 per cent purchased steel scrap. Carbon is picked up by the metal from the coke. The coke-to-metal ratio will vary, depending on cupola operation; however, ratios varying from 1:6 to 1:8 are most common. The quantity of limestone normally used ranges from 10 to 25 per cent of the coke charge. In all-steel charges the limestone used is from 8 to 12 per cent of the coke charged. Some operators make an addition of fused soda ash instead of the limestone. The charge is so balanced that the metal leaving the cupola will have an analysis in the following range:

	Per Cent
Carbon	2.75–3.50
Silicon	1.30–2.00 (depending on temperature required)
Manganese	0.50–0.60
Phosphorus	normally not over 0.04
Sulphur	normally not over 0.10 (as tapped)

The temperature of the metal coming from the cupola should preferably be above 2700°F. A number of operators report temperatures of about 2850°F. as their average temperature. The temperature of the iron going into the converter will be between 2730 and 2650°F. By controlling the silicon content of the cupola metal, the temperature of the steel at the end of the blow can be regulated between 2900 and 3200°F. or higher if desired. One foundry reports a temperature of 3450°F. for steel coming out of the converter for pouring by shanking methods.

The sulphur content of the molten iron coming from the cupola prior to soda treatment is always higher than the sulphur content of the metal charged to the cupola because some sulphur is picked up through contact with the coke.

The metal as it comes from the cupola is very carefully skimmed of

slag by mechanical means. It is quite important that all the slag is removed from the cupola metal since it cuts down the efficiency of the desulphurizer. The removal of part of the sulphur is accomplished by treating the molten iron with (1) soda ash (sodium carbonate, Na_2CO_3), (2) caustic soda (sodium hydroxide, $NaOH$), and (3) soda-ash caustic mixtures. These chemicals are rated in commercial parlance on the basis of their sodium oxide (Na_2O) content. Thus caustic soda with 76 per cent Na_2O is proportionately more active than soda ash with 58 per cent Na_2O. Under favorable conditions upward of 85 per cent of the sulphur present in cupola iron within the range of from 0.06 to 0.13 per cent sulphur can be removed with caustic soda in one treatment. The maximum reduction obtainable when treating similar irons with soda ash is about 75 per cent.

The iron from the cupola is tapped into the ladle until it is partially filled; then some of the required alkali is added. During the time the ladle is being filled, the remaining portion of the desulphurizer is added. A violent boiling action takes place in the ladle and a highly fluid slag is formed. The reaction is allowed to continue until the boiling action ceases, and then the slag, which has absorbed sulphur from the metal, is removed as completely as possible. The metal is then ready for transfer to the converter.

Starting with hot, clean cupola iron and with good desulphurizing practice, the per cent sulphur reduction with fused soda ash will vary about as shown[33] in Table XXXIII.

TABLE XXXIII.—PER CENT SULPHUR REDUCTION WITH FUSED SODA ASH
Evans[33]

Sulphur in iron, original per cent	Fused soda ash, lb. per ton						Maximum reduction, per cent
	1.2	5	10	20	30	40	
	Sulphur in iron, final per cent						
0.15	0.13	0.11	0.07	0.055	0.045	0.040	73
0.12	0.11	0.09	0.06	0.048	0.042	0.036	70
0.09	0.085	0.075	0.055	0.045	0.037	0.032	65
0.07	0.07	0.06	0.05	0.04	0.032	0.028	60

Erratic or unsatisfactory results with desulphurizing, sometimes reported by foundry operators, usually have originated in contamination of the metal by the slag or any siliceous impurities covering the metal and preventing rapid and intimate contact between the metal and soda slag.

The following procedure is recommended by Evans[33] in ensuring both clean iron and maximum contact between the slag and metal: The cupola taphole should be smaller than average, so as to extend the tapping

time over a longer period and thus to prolong the stirring action by the stream of iron. A small bed of iron should always be held back in the cupola at the end of the tap to prevent any cupola slag from coming out with the last of the iron. For maximum desulphurization for converter practice, the use of pigs or lumps of fused soda ash (which penetrate the slag and melt in contact with the metal, gradually releasing fresh alkali) usually gives lower and more uniform sulphur.

Converter Practice.—Following desulphurization, the cupola metal is accurately weighed, carefully skimmed of all desulphurizing slag, and placed in the converter, which has been turned down to the charging position. The converter is then turned up until the surface of the metal is level with the bottom of the tuyères as observed through the wind-box door. The converter is then at an angle of between 4 and 10 deg. as indicated by the pointer mounted on one of the trunnions. In general, this will be about 8 deg. forward. One foundry claims that if the blowing angle is 13 deg., the blowing time is cut by 6 to 8 min.

Care should be exercised to add or remove cupola metal so that the selected tilt can be maintained. The inside of the converter will increase in size, owing to lining erosion, as converter operations proceed. Thus, in order to hold the proper angle, the weight of the converter charge will increase. Charts may be prepared such as those of Kain and Sanders[36] in Table XXXIV, which show the increase in weight of charge necessary as wear of converter lining takes place.

TABLE XXXIV.—TOTAL CONVERTER-CHARGE WEIGHT TO COMPENSATE FOR LINING
EROSION
Kain and Sanders[36]

Days	Total Converter-charge Weight, Lb.
Monday	2,700
Tuesday	2,700
Wednesday	2,800
Thursday	3,000
Friday	3,000
Saturday	3,200

The wind-box backing plate is clamped in place. The operator notes the angle and rotates the converter a few degrees forward, to prevent the metal from surging back into the tuyères as he turns on the blast pressure. A pressure of 3 to 5 lb. is usually maintained. The operator then rotates the vessel back to the previously noted position and records the time; the blow has started.

If the vessel lining and the metal in the cupola are quite hot, a small pale-yellow flame should come up almost immediately, accompanied by a large volume of smoke and pale-yellow sparks. During the first period of the blow the flame is not very clear. Within approximately 3 to 5

min., the next period, known as the "manganese period," appears. The flame becomes intensely white with sharply defined outlines. The flame attains its full height in about 3 min. and then decreases. About 3 min. later the flame again ascends as the metal starts to boil vigorously, causing some slag to be emitted. The pressure is then reduced slowly to about 2 lb. By this time the bulk of the silicon and manganese has been oxidized. In some cases it is the practice to reduce the pressure only momentarily during the boil, and the pressure is put on again after the boil.

During the third period, the carbon period, the flame is not so clearly defined. It becomes flickery with feathery edges and a yellowish-white
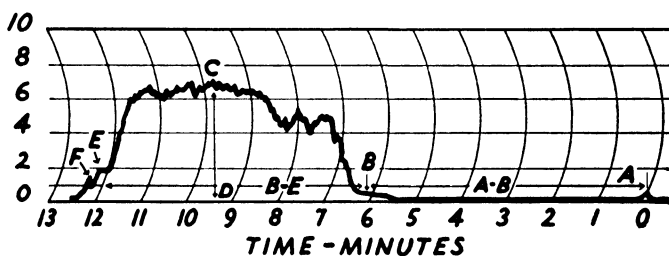
A	START	C-D	HEIGHT OF CURVE
A-B	SILICON BLOW	E	END POINT
B-E	CARBON BLOW	F	FLASHBACK

Fig. 23.—Significance of the photoelectric-cell flame curve. (*Work.*[34])

color of normal size. As the oxidation of carbon proceeds, the flame ascends until it is the longest of the entire blow; it is quite intensive in brightness and has feathering edges. As the carbon becomes exhausted, the flame becomes streaked and diminishes in length until it drops. The blast is shut off, and the converter is turned down. To obtain a hot blow, ferrosilicon is sometimes thrown into the vessel directly after the appearance of the first flame.

The bath at the end of the blow will contain approximately 0.05 per cent carbon, 0.02 to 0.05 per cent silicon, and from 0.01 to 0.03 per cent manganese. About 10 to 15 per cent of the metal will be lost in oxidation.

The flame-illuminosity characteristics can be more readily observed if the photoelectric-cell apparatus, developed by the Jones and Laughlin Steel Corporation, is used. A typical curve and its interpretation[34] are illustrated in Fig. 23. The photoelectric cell is directly dependent on the intensity of the flame. This intensity varies considerably with

1. The temperature of the metal in converter
2. The silicon content of the metal
3. The size of the nose of the converter

The third item is most important.   As the nose enlarges, as a result of erosion from the blows, the flame becomes larger, and adjustments must be made to keep the recorded results on the chart.

The time required for the actual blowing operation will vary somewhat under different conditions but is usually from 12 to 16 min.   As soon as the blow is completed and the converter turned down, deoxidizers are added to the bath.   A convenient method for making these additions is by the use of silicomanganese.   Some operators, however, desire to make an 80 per cent ferromanganese addition as the first deoxidizer used.   The ferrosilicon addition will then be made to the ladle after the recarburizer is added to the bath.   Other operators prefer the ferrosilicon addition to precede the ferromanganese addition and both to be added to the converter.   The addition is made by pushing back the slag and throwing in the alloys followed by rabbling with a bar or by rocking the converter back and forth.   It is the experience of some operators that by placing the ferromanganese in the bottom of the ladle and tapping the steel upon it, a higher recovery of manganese is obtained.

The next addition is the recarburizing metal, which is taken directly from the cupola.   The amount of cupola metal added is regulated by the carbon content desired and should be weighed accurately.   The bath is again stirred or rabbled and then tapped into a waiting ladle.   Aluminum in quantities of from 1 to 3 lb. per ton is added to the steel in the ladle.

It is customary to use a slag skimmer in the mouth of the vessel, to hold back the slag until the ladle is almost filled.   If there is much slag in the vessel, it may be partly skimmed out before tapping.   After the metal is tapped, the converter is turned down completely, to drain off any slag remaining in the vessel.   An inspection is then made of the lining and the tuyères, and repairs are made if necessary.   The converter is now ready to receive metal for the next blow.   The total cycle of charging the converter, blowing, deoxidizing, tapping, and slagging requires about 25 to 30 min., although the blowing time consumes only from 12 to 16 min.

Low-alloy steels are being made by the converter process.   Nickel is often added by introducing it into the stream of the cast iron as it leaves the cupola.   In other cases nickel and molybdenum are added to the converter.   The remaining alloys such as chromium, vanadium, copper, and sometimes molybdenum are introduced by ladle additions after the blow.   Very little converter steel is made, however, with solid additions containing more than about 3 per cent alloy.

A few suggestions regarding operations in general should be added. The importance of early first-light flame is great.   It is believed that the longer the blow takes, the colder the steel will become.   This first light may be expedited in a number of ways; each method is primarily one of breaking the slag crust that forms over the metal.   One way is to throw

in a lump of about 4 lb. of ferrosilicon, or a casting of about 4 lb., or to roll the converter forward about 10 deg. while blowing and then back to the original position. Very hot cupola metal is the most important factor in blowing steel, particularly when a shanking casting practice is to be followed.

Triplex Method.—The triplex method has been developed for the continuous production of steel castings. This method involves the use of the cupola, the converter, and the electric furnace. The development of this process was the result of a demand for an alloy cast steel of carbon content of 1.50 per cent. The triplex method was modified to produce cast steel ranging from 0.25 to 0.50 per cent carbon, as a result of the increased production of steel castings required for the Second World War. The cupola charge for the triplex method usually consists of steel scrap plus silicon-bearing material, such as 50 per cent ferrosilicon, silicon briquettes, or pig iron in such proportions that the silicon content of the cupola metal will be from 1.75 to 2.00 per cent. The cupola metal is desulphurized in the ladle prior to blowing in the converter. Also, the ladle treatment for the removal of phosphorus after blowing may be applied.

In a number of cases[38] the steel is deoxidized and recarburized by additions of cupola metal, ferromanganese, ferrosilicon, and aluminum, before it is delivered to the electric furnace. The electric furnace in this case operates as a holding furnace and is lined with silica refractories. The composition and temperature of the steel are adjusted in the electric furnace; and if alloys are required, it is convenient to add them here. It is customary to have about a half hour's supply of steel in the electric furnace at all times.

It is reported by Gregg[38] that, when this process was first put into operation, some difficulty was experienced in maintaining a neutral nonoxidizing slag. Such a type of slag is highly essential because an oxidizing slag will result in a loss of carbon, silicon, and manganese.

The carbon content of the electric-furnace bath is checked after each addition of a converter heat, so that proper adjustments may be made to the bath to provide for uniform metal analysis in the practice of continuous pouring. A practice of continuous pouring has been used to advantage in the highly mechanized malleable- and gray-iron foundries producing automotive castings. This scheme of operations is now being extended to the production of steel castings in which conveyer pouring mechanisms are used.

It is not the practice in all cases to deoxidize the steel as it comes from the converter. Two operators of the triplex process tap the converter steel into the electric furnace without deoxidizing it. The carbon content is adjusted in the electric furnace, but all deoxidizers are added to the

ladle when a portion of the electric-furnace bath is drawn off for the continuous pouring operations.   The reason for this is the belief that there is considerable gas pickup by the metal in the electric furnace if the converter metal is deoxidized before adding it to the electric furnace. This has been the experience of electric-furnace operators whenever they hold deoxidized steel in the electric furnace.

There are several variations of the triplex process that are worthy of consideration.   It has been suggested that the electric furnace be basic lined.   If this were done, it would be possible to handle cupola metal of higher phosphorus content than 0.04 per cent, such as might be obtained in melting cast-iron scrap.   The phosphorus could then be removed by basic oxidizing conditions and the removal of the high-phosphorus slag at stated intervals.

Another and perhaps more promising method of operation involves the use of the dephosphorizing method developed by Yocom.[39]   In this process care is taken to keep the converter slag from entering the ladle with the blown steel.   The dephosphorizing material is added to the stream of metal as it is being poured from the converter into the ladle. The mixture used generally consists of about 50 per cent impure lime (calsifer), 30 per cent roll scale (dried), and 20 per cent flux (fluorspar). This mixture is added in the cold state.   Such a mixture, when melted together to form a slag, analyzes about as follows: CaO, 48 per cent; Fe_2O_3, 28 per cent; SiO_2, 7 per cent; Al_2O_3, 7 per cent.

Ferromanganese and the mixture are added concurrently at first and then are followed by the mixture alone.   The addition is complete by the time the ladle is a little more than half full.   A violent reaction takes place in the ladle.   From 60 to 80 lb. per ton of dephosphorizing material is used for 6½-ton heats for different grades of steel.   The operation is complete in about 30 sec., and the phosphorus is reduced from about 0.100 per cent to 0.020 to 0.040 per cent.

It has been stated by several operators of 1- to 2-ton side-blow converters that this method does not work so effectively with small heats as with the larger Bessemer converters, and the undesirable loss of temperature during the treatment is pronounced.   If, however, the Yocom method proves feasible for phosphorus removal in small converter heats and the metal is transferred to a basic-electric furnace, it will be possible to operate with a single refining slag, which will allow further reduction of the sulphur content.

Development is now going forward in providing basic linings for the cupola for a distance of about 2 ft. above the tuyères.   Reports show that this practice has great possibilities as to phosphorus removal.   It would then be quite possible that the dephosphorized steel could be tapped into an acid-lined electric furnace, providing the dephosphorizing

slag is very carefully removed from the ladle prior to pouring the metal into the electric furnace. This manipulation may be quite difficult to carry out, since the dephosphorizing slag is an extremely fluid, thin slag. It can be done, however, by employing a suitable skimming practice.

The advantage of the triplex process is that there is a large saving in power costs, since the electric furnace, when operated as a holding unit, consumes only about 50 to 75 kwhr. per ton with acid operation. Although the triplex process is a recent development, it is in successful operation today in several foundries and is being seriously considered for use in others. In one case, a manufacturer of steel castings has been able, through the triplex process, to obtain a production four times the normal rated capacity of an electric furnace.

REACTIONS IN THE ACID PRACTICE

Carbon is eliminated from the steel bath by the reaction

$$C + FeO = CO + Fe$$
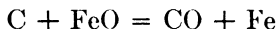
The equilibrium and the reaction-rate constants of this reaction do not depend upon the temperature so much as upon the total concentration of carbon in the melt. The minimum content of iron oxide for a given concentration of carbon in the melt will be attained when the rate of carbon elimination is approaching zero. Even though the reaction rate does not depend entirely upon the temperature, the reaction does not take place with any vigor until the bath attains a high temperature.

The factors affecting the oxidation of the metal are: (1) the carbon content, (2) the slag composition and viscosity, and (3) the temperature. As the carbon content of the metal drops, the FeO content of the metal increases. The manner in which this may take place in an acid open-hearth heat is shown in Fig. 24. In the acid practice a drop in the FeO content of the metal is noted before the addition of the deoxidizers. This drop is caused by the change in slag composition and viscosity. As the slag picks up more and more silica from the furnace lining, the free oxide content of the slag decreases and the viscosity of the slag increases. Both of these factors tend to decrease the rate of diffusion of iron oxide from slag to metal; and since the carbon reaction is proceeding at all times, there is necessarily either a decrease in the rate of pickup of iron oxide or an actual decrease in the iron oxide content of the bath.

One of the main reasons that the acid practice has gained wide use in the steel casting industry is the simplicity of the control of iron oxide. In Fig. 25 is shown the relation between the FeO in the slag and metal for all ratios of SiO_2, $CaO + MgO$, and FeO in the slag. In the upper corner, above the heavy line, are the acid slags. Each of the curved lines in the diagram represents a definite relationship between the iron oxide in

the slag and that in the metal (the partition constant).   Only one of
these lines extends into the acid region—the 0.0020 line at 2900°F.   This
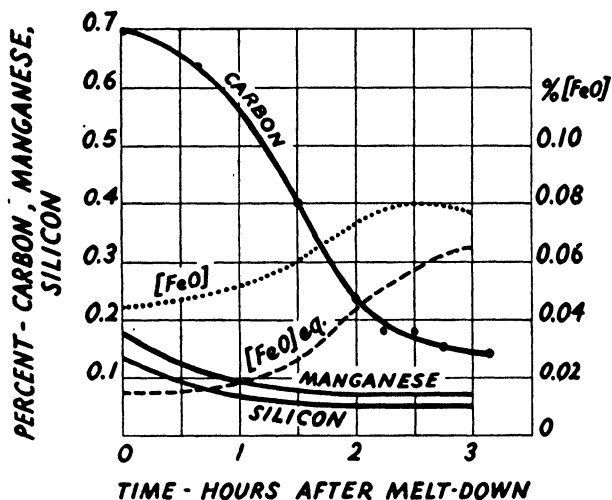
Fig. 24.—Increase in FeO content as the carbon drops in an acid open-hearth heat.

means that for a single temperature, there is only one ratio of iron oxide
in the metal and in the slag, for the acid practice.
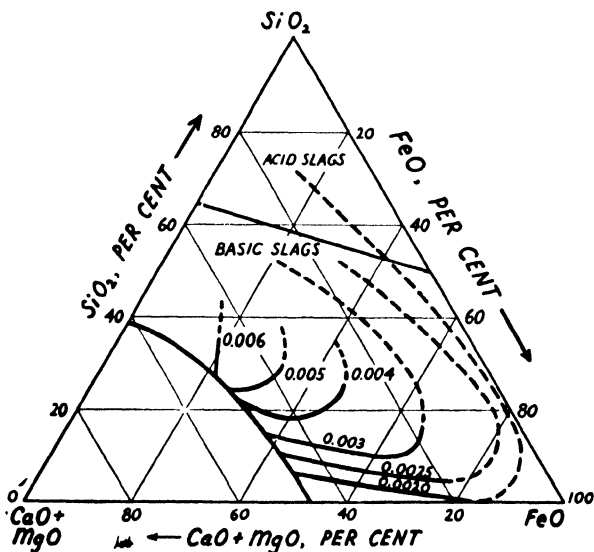
Fig. 25.—Influence of acid-base ratio on partition constant at 2900°F.   (*Fetters and Chipman.*[41])

A further important item is that the partition constant with acid slags
is low.   For example, metal under an acid slag of 10 per cent FeO at

2900°F. will contain 0.090 per cent FeO. To obtain equally low FeO in the metal at 2900°F. under a normal basic slag, the FeO content of the basic slag must be less than half of the FeO content of an acid slag, or 4 per cent FeO or less.

The partition constant for acid slags is considerably influenced by temperature (Fig. 26). The reason for the difference in the data of the chart is that the iron oxide was determined by two methods: (1) carbon monoxide by vacuum fusion (solid line), and (2) as alumina, Al_2O_3 (dotted line). Attention should be given to the influence of temperature on the partition constant, for when steel is deoxidized under an acid slag at
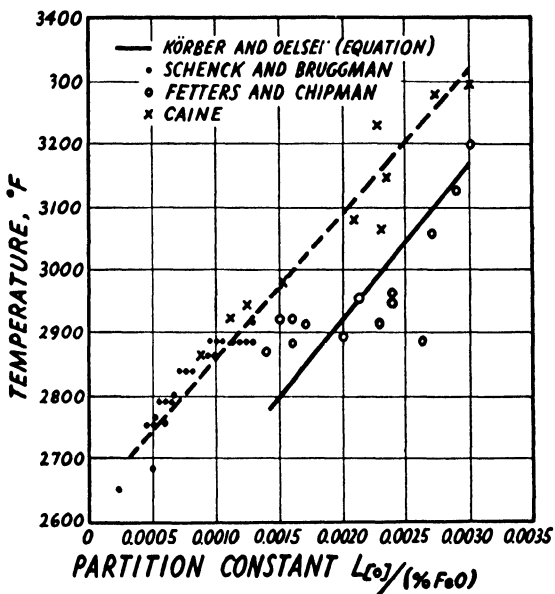
FIG. 26.—Influence of temperature on partition constant for acid slags. (*Caine.*[41])

3100°F., there is nearly 70 per cent more FeO in the steel than if the deoxidation took place at 2900°F. The higher iron oxide content of the metal will require a proportionately greater amount of deoxidizer.

Acid slag represents a solution of oxides of iron, manganese, and silicon, which are present partly free and partly in the combined state. All that can be said at present is that the amount of free oxides of iron and manganese represents only a small fraction of the total analytical concentration of these oxides in acid slag.

The acid slag is at all times saturated with silica so long as the temperature is constant; the silica content of the slag that is in contact with the acid hearth tends also to remain constant. The quantity of silica present will depend upon the temperature. The MnO content of the slag is largely in a state of combination with the silica.

The addition of a base such as CaO to the slag upsets the equilibrium of SiO_2-FeO-MnO and gives to the slag a temporary excess of MnO and FeO, which oxidizes the carbon content of the bath and thereby lowers slightly the MnO and FeO content of the bath.

Under a slag consisting of a mixture of manganese and iron oxides, the manganese and silicon content of the metal is oxidized according to the following reactions:

$$Mn + Fe = MnO + Fe$$
$$Si + 2FeO = SiO_2 + 2Fe$$
$$Si + 2MnO = SiO_2 + 2Mn$$
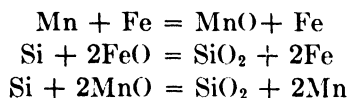
These reactions take place at low concentrations of carbon or at low temperatures; hence they occur early in the melting period and late in the oxidizing period. The greater part of the oxidation takes place during the melting period.

Körber and Oelsen[40] noted that the content of silicon in the metal increases very rapidly with the increase of manganese concentration. At a manganese content of over 0.4 per cent, the relative changes in slag composition are very small, whereas the relative changes in the content of silicon are considerable. Thus the reaction

$$SiO_2 + 2Mn = Si + 2MnO$$
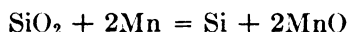
is of predominant importance in determining the reduction of silicon to the metal from the slag. Silicon reduced in this way from the slag or furnace hearth will not be accompanied by inclusions. The FeO content of the metal bath will decrease as the silicon is reduced from the slag.

In the electric furnace at temperature of approximately 3100°F., silicon is reduced from the slag by the following reaction:

$$SiO_2 + 2C \rightarrow Si + 2CO$$
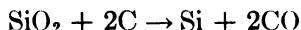
The evolution of CO gas produces an active bath that usually appears as a carbon boil. The greater the manganese content of the bath, the greater is the amount of silicon reduced.

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GASES, DEOXIDATION, POROSITY, AND INCLUSIONS

GASES

Solubility of Gases. Oxygen.—It is customary to express the solubility of oxygen (O_2) in iron or steel as the solubility of iron oxide (FeO). The solubility of oxygen in liquid iron was determined by Tritton and Hanson[1] to be 0.21 per cent oxygen in iron at the melting points, 1535°C. (2795°F.), which is equivalent to 0.94 per cent FeO. Herty and Gaines[2] checked the results of Tritton and Hanson and also showed that the solubility increases with temperature as a straight-line function until 1700°C. (3090°F.), where 2 per cent FeO is present. Körber's[3] results agree quite well with those of Herty and Gaines.

Chipman and Fetters[4] made a very complete review and study of the oxygen solubility in liquid iron, and their results are shown in Fig. 27. These values apply to nearly pure iron with a slag consisting principally of FeO. Such concentrations would never be reached in practice. Actually, the amount of FeO in a steel bath is dependent upon the FeO content of the slag. The iron oxide tends to divide itself between the slag and metal according to a distribution ratio that changes with slag composition and temperature.

The solid solubility of oxygen in iron is, according to Krings and Kempkens,[5] 0.11 per cent (0.49 per cent FeO) at 715°C. (1319°F.). However, Ziegler[6] states that the solubility of oxygen in solid iron is negligible below the critical temperature; but that above 900°C. (1650°F.) it increases rapidly with temperature and reaches a maximum at 1000°C. (1830°F.) of 0.10 per cent, corresponding to 0.45 per cent FeO.

In all steelmaking processes iron oxide is dissolved in the steel at some stage of the melting. In the basic-electric furnace the amount present during the latter part of the refining period is at a minimum, whereas in the open-hearth furnace and in the converter there may be as high as 0.50 to 0.60 per cent FeO dissolved in the molten steel.[7] The amount of oxygen in molten steel is tremendously affected by deoxidizers; upon deoxidation the amount of iron oxide is reduced, depending upon the type and amount of deoxidizers used. Some oxygen will always be present as oxidizers of foreign metals and metalloids, as MnO, Al_2O_3, and SiO_2. Oxygen in these forms, however, is not to be considered as alloyed with the iron, since it is present as nonmetallic inclusions. A representa-

tive value of oxygen in low-oxygen steels is about 0.007 per cent; steels of the high-oxygen type contain perhaps 0.040 to 0.060 per cent.   Other types have intermediate values, from 0.010 to 0.030 per cent oxygen.[8]

The presence of oxygen is believed to affect the solubility of carbon in austenite and ferrite and thereby to affect the structure of the iron or steel, though this is not certain.   Such properties as increased strength and hardness; lower ductility, elastic limit, impact resistance, magnetic permeability; tendency towards caustic embrittlement, irregularity in carburization, certain corrosion phenomena; the presence of flakes, seams, bands, ghost lines and hair cracks; and the tendency to age-harden, all have been attributed from time to time to the presence of oxygen.
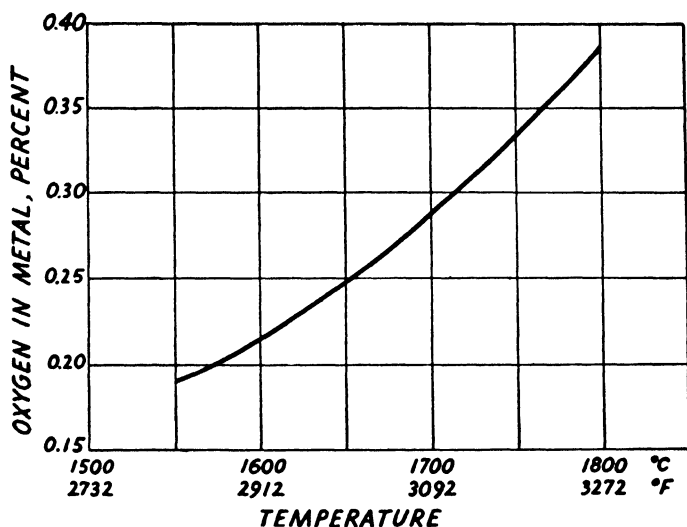
Fig. 27.—Effect of temperature upon oxygen content of liquid ingot iron.   (*Chipman and Fetters.*[4])

Carbon Monoxide and Carbon Dioxide.—Carbon monoxide and carbon dioxide are always found in gases evolved from steel, either when solidifying or when subjected to vacuum treatment.   Klenger[9] concluded, after considerable study, that CO and CO_2 were merely products of the reaction $FeO + C \rightarrow Fe + CO$.   A discussion of this reaction and its relation to the formation of porous castings will be discussed later in greater detail.   The work of Kinzel and Egan[10] establishes the fact that CO and CO_2 are not appreciably soluble in molten steel and are not present in steel as such.

Hydrogen.—The solubility of hydrogen (H_2) in steel at 1 atmosphere pressure was established by Sieverts.[11]   Figure 28 shows the hydrogen solubility at various temperatures.   These results have been carefully checked,[12] and it has been established conclusively that hydrogen is

dissolved in steel atomically. The quantity of hydrogen that will dissolve in steel is directly proportional to the square root of the pressure.[13]

It is evident from Fig. 28 that the sharp decrease in solubility with decreasing temperature will result in a steel with a high degree of supersaturation. The greater portion of the gas held by the liquid metal tends to be lost during solidification. It has been estimated that the loss of three volumes of hydrogen, measured at the temperature of the melt, per volume of metal is not unlikely even for normal castings. However, to hold the excess gas in solution in the metal, Zapffe and Sims[14] show
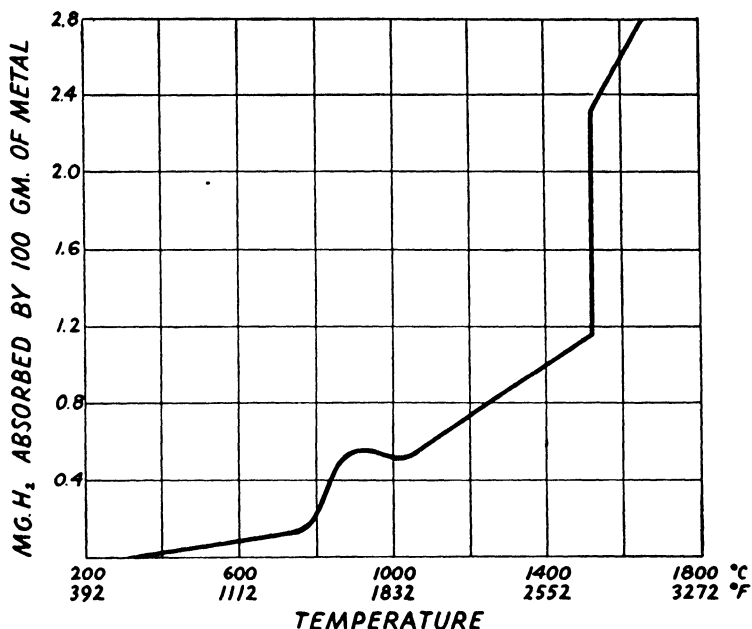
Fig. 28.—The effect of temperature on the solubility of hydrogen in iron. (*Sieverts.*[11])

that greatly increased pressures of hydrogen outside the metal are required. Since the decrease in solubility is considerably more than one thousandfold between the temperature of liquid steel and room temperature, over a thousandfold increase in the partial pressure of atomic hydrogen would be required to prevent formation of hydrogen from a cold steel that had been saturated previously when molten.

In Fig. 29 Zapffe and Sims[15] show that for a given temperature and pressure, nickel is a better solvent for hydrogen than is iron. Alloys of iron and nickel, in general, absorb amounts of hydrogen roughly proportional to the composition of the alloy.[16,17] Solubility measurements of hydrogen in manganese[16] show values several fold higher than those of iron. The Cr-H curve,[16] on the other hand, indicates that chromium is a poorer solvent for hydrogen than is iron. Gassiness, the degree

of which is a rough measure of the change in solubility from liquid steel to solid, should therefore tend to be suppressed by alloying with nickel or manganese, since these alloys hold more of the gas in solution, whereas chromium should exert the opposite effect because of a decreased solubility.

A popular belief often expressed is that deoxidizers, such as silicon, markedly increase the solubility of steel for hydrogen. Zapffe and Sims[15] point out that this is not true and that such a theory is not necessary to explain the phenomenon of porosity. The addition of deoxidizers results in the decrease of the oxygen activity of the steel, and in so doing
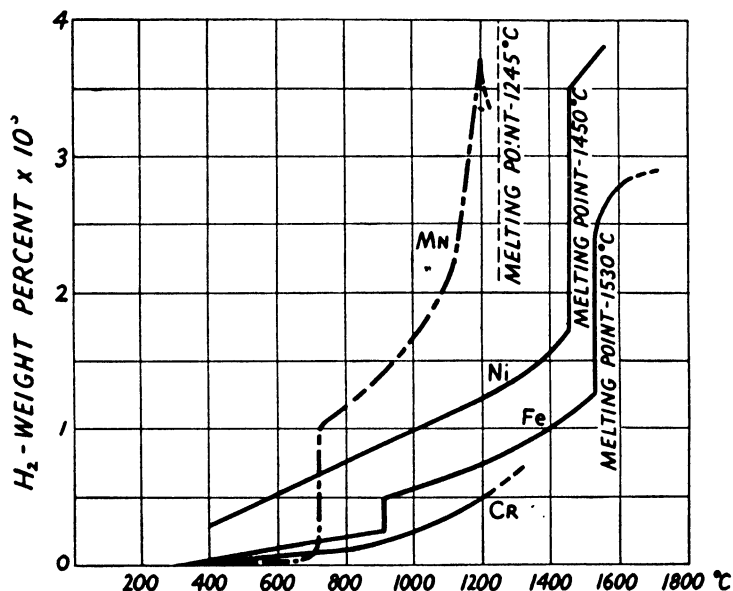
FIG. 29.—Absorption of hydrogen by iron, nickel, chromium, and manganese from hydrogen at a pressure of 1 atm. (*Zapffe and Sims.*[15])

they increase the capacity of the steel for absorbing hydrogen in accordance with the equation for the hydrogen-oxygen equilibrium. Thus oxygen has a great influence on the hydrogen solubility in steel and also on the action of the hydrogen that is retained.[18] This latter influence works through the reversible reaction $H_2 + FeO \rightleftharpoons Fe + H_2O$. Hydrogen gas may be used to deoxidize steel, though not very effectively and not without saturating the steel with hydrogen. On the other hand, oxygen can remove excessive amounts of hydrogen. Neither can entirely eliminate the other.

Hydrogen has a strong tendency to form a solid solution rather than to precipitate as bubbles during freezing. Sims and Zapffe[22] conclude that hydrogen dissolved in steel, even to saturation under one atmosphere

of H_2, is unable to initiate bubble formation.   If a bubble of any other gas is introduced under such conditions, hydrogen will diffuse into it and cause it to grow with great rapidity.

The phenomenon of bubble formation in the case of hydrogen in liquid iron therefore comprises two factors:[15]

1. A concentration factor, which immediately demands at any given temperature that a particular pressure of molecular hydrogen be present at all surfaces of the liquid within or without; and

2. A nucleating agent that will create the surface within the liquid that can lead then to the collection of hydrogen and consequent bubbling.

TABLE XXXV.—GAS CONTENT OF STEELS—PER CENT BY WEIGHT
Motok[19]

Grade	Oxygen	Hydrogen	Nitrogen
	Open-hearth Steels		
SAE 2015	0.0065 –0.0257	0.00048–0.00195	0.0029–0.0069
SAE 2340	0.0067 –0.0098	0.0006 –0.0024	0.0032–0.0050
SAE 4615	0.0011 –0.1021	0.0001 –0.0033	0.0020–0.0194
2¾ Ni-Mo	0.0010 –0.0104	0.00005–0.00131	0.0020–0.0054
	Electric-furnace Steels		
SAE 1035	0.0108 –0.0110	0.00036–0.00041	0.0077–0.0113
Amola	0.0044 –0.0066	0.00007–0.00031	0.0079–0.0133
SAE 4615	0.0083 –0.0305	0.0005 –0.0024	0.0062–0.0146
SAE 3115	0.0008 –0.0181	0.00026–0.00127	0.0063–0.0157
SAE 4320	0.0020 –0.0077	0.00054–0.00057	0.0066–0.0069
SAE 6130	0.0056 –0.0152	0.00016–0.00037	0.0051–0.0140
Cr-Mn	0.0123 –0.023	0.00008–0.00017	0.0089–0.0143
SAE 52100	0.0112 –0.0181	0.00083–0.00183	0.0066–0.0106
C-105	0.003 –0.113	0.00009–0.0182	0.0012–0.0322
SAE 9260	0.0042 –0.0063	0.00015–0.00024	0.0068–0.0078
SAE 9260 plus Mo	0.0050 –0.0070	0.00026–0.0003	0.0074–0.0094
18-8	0.01754–0.0322	0.00026–0.00141	0.0324–0.0414
18-8 Cb	0.0164 –0.0235	0.00070–0.0020	0.0310–0.0381
HC stainless	0.0077 –0.0610	0.00050–0.0007	0.1879–0.2050
Mn-Cr stainless	0.018 –0.023	0.0009 –0.0010	0.0607–0.097

Hydrogen is usually present in negligible amounts in steel at atmospheric temperature.   If present in larger amounts, hydrogen may affect the properties of the metal to a noticeable extent.   The brittleness that follows certain types of pickling or the embrittlement due to flakes, shatter cracks, etc., has been attributed to the presence of hydrogen.

The hydrogen content of solid steel will vary from 0.00005 to 0.018 per cent by weight. On the whole, the electric-furnace steels contain more hydrogen than the open-hearth steels. An average hydrogen content is approximately 0.00065 per cent. Table XXXV shows the variations in hydrogen, oxygen, and nitrogen that have been obtained in different classes of steels.

Nitrogen.—The solubility of nitrogen in iron at the melting point, under approximately 1 atmosphere of nitrogen, has been determined by Chipman and Murphy[20] to be 0.39 per cent by weight. They also found that the presence of silicon and aluminum did not affect the solubility; alloying elements, when present in sufficient amounts, obviously change the value. This value is above that encountered as a result of the absorption of nitrogen from air or from flushing steel with nitrogen.[15] The maximum nitrogen content that can be obtained in solid iron-nitrogen alloys is 11.3 per cent, as observed by Hägg.[21] The maximum solubility of nitrogen in alpha iron is 0.42 per cent at 591°C. (1096°F.) and 0.015 per cent at room temperature.

Commercial steels contain nitrogen in quantities shown in Table XXXV. Bessemer and electric-furnace steels are likely to contain the most nitrogen, in a range up to 0.10 or 0.20 per cent. Open-hearth steels contain from 0.003 to 0.008 per cent nitrogen. The nitrogen absorbed by molten steel forms a stable nitride that is not easily decomposed.[18] Nitrogen, like hydrogen, does not fully evolve or bubble inside molten steel. High nitrogen contents may be produced by the addition of high-nitrogen alloys to the melt.[15] Steels prepared in this way may contain from 0.055 to 0.07 per cent nitrogen after effervescence of nitrogen from these steels is complete.

EFFECT OF DEOXIDATION

The deoxidation of steel refers, strictly speaking, to the removal of oxygen from the steel. Oxygen is not removed from the steel, however, but is merely fixed in a less active condition. The term "deoxidation" is used in a broader sense and includes all the phenomena in the final preparation of steel that would permit it to solidify free from gas cavities and from dissolved oxygen of all kinds.

Steel that is ready for casting into steel castings is dead-killed steel. It is prepared as such by refining and by deoxidation practice. The deoxidation reactions involved are

$$FeO + C \rightarrow Fe + CO \qquad (1)$$
$$FeO + Mn \rightarrow Fe + MnO \qquad (2)$$
$$2FeO + Si \rightarrow 2Fe + SiO_2 \qquad (3)$$
$$3FeO + 2Al \rightarrow 3Fe + Al_2O_3 \qquad (4)$$
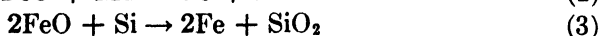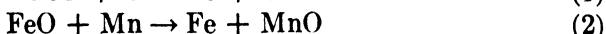
Equation (1) is the reaction of decarburization, which produces the carbon boil. The equilibrium constant for this reaction

$$K = (\text{per cent FeO}) \times (\text{per cent C})$$

is between 0.0005 and 0.005. The boil often ceases when the product is as high as 0.03, or when the FeO present is from 6 to 60 times as much as would be possible under equilibrium conditions for a given carbon content.

As long as an equilibrium exists between C and FeO, there will be no reaction and no evolution of CO gas. If steel is solidified as it exists at the end of the oxidation period, equilibrium conditions are upset and CO gas is liberated. During cooling and solidification the solubility for FeO is greatly decreased, and the deoxidation constant decreases; this results in a great increase in its activity. Such an activity is featured by rimming steel.

The CO gas is not released in steel for castings upon solidification because the quantity of FeO is decreased during the refining period and is further decreased by the use of deoxidizers just prior to tapping the heat. During the refining period, Eq. (1) proceeds toward an equilibrium value depending upon the carbon content and upon the temperature of the bath.

The FeO content is further decreased by the action of the deoxidizers according to the reactions (2), (3), and (4). It is shown by Hultgren and Phragmen[23] that as little as 0.05 per cent silicon of 0.5 per cent manganese will considerably affect the rimming[1] characteristics of a steel. It is improbable, from an examination of the published data on equilibrium constants by Chipman[24] (Fig. 30), that CO gas would be liberated in accordance with reaction (1) in a killed cast steel containing 0.25 per cent silicon or greater and 0.50 per cent manganese or greater. Sims and Zapffe[18] give a good example on this point, as follows:

A cast steel containing 0.25 per cent carbon and 0.30 per cent silicon at a temperature of 2900°F. will have both of these elements in equilibrium with an FeO content of approximately 0.03 per cent. Just above the freezing point, the FeO would have to be above 0.02 per cent to react with the carbon, but only 0.01 per cent FeO can exist in the presence of the 0.3 per cent silicon. The residual silicon serves as a reservoir that will react with any extra FeO that might be introduced, each unit of silicon reacting with five times its weight per cent FeO. To produce a C-FeO reaction, about 0.10 per cent silicon would have to be oxidized, and this would require the introduction of 0.50 per cent FeO.

The silicon-deoxidation reaction reaches equilibrium apparently almost instantly—in a matter of seconds. If, however, the iron oxide

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[1] Rimming—the evolution of CO during the solidification of steel.

of the slag is high (over 15 per cent) before the addition of silicon, there will be a slow reaction between the metal and the slag, oxidizing the silicon in the metal and reducing the iron oxide in the slag until a balance is attained between the metal and slag. This reaction takes place only at the slag-metal interface, and hence it requires considerable time to reach an equilibrium condition. However, the moment that a tilting furnace is tapped, an intimate mixture of slag and steel is obtained and the slag-metal interface reaction is speeded up and equilibrium conditions are attained in a matter of seconds. The silicon added for the purpose of deoxidizing the steel would also react to lower the FeO content of the slag.

% ALLOY REMAINING IN STEEL
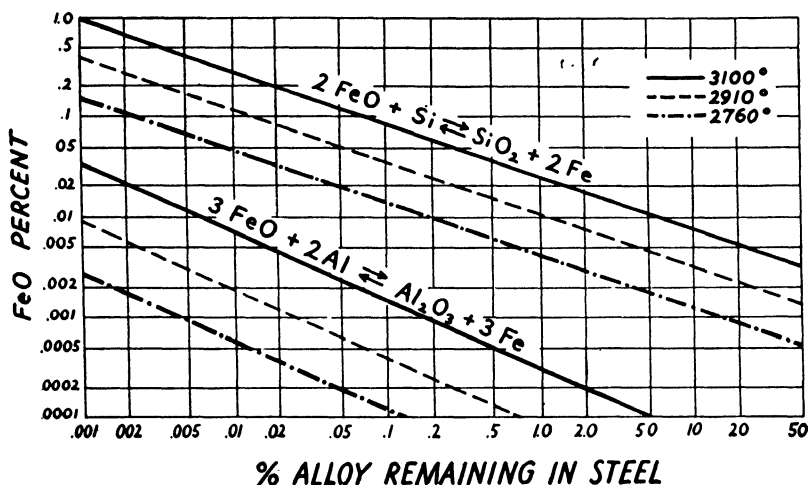
Fig. 30.—Relationship between silicon, aluminum, and FeO.

Silicon is apparently a good deoxidizer and, if present in sufficient quantities, it is quite adequate to prevent the formation of CO. It has been pointed out time and again by Sims[25] that for a medium-carbon steel, the minimum silicon content that will suppress the reaction to form CO gas is about 0.25 per cent. For safety, that figure usually is given a margin of 10 points, or a total content of 0.35 per cent silicon. All silicon additions up to this value rapidly lower the FeO content, but beyond this value the returns are small. It is probable that CO can be discounted entirely as a primary factor in causing porosity, considering the silicon contents normally used in steel castings. Nevertheless, when gas evolution is once started from hydrogen- or nitrogen-source conditions, the evolved gases most certainly contain CO.

Manganese is considered a deoxidizer; but if dependence is placed upon manganese alone, it would be impossible to obtain a killed steel. Manganese has a higher affinity for oxygen than iron, but in amounts

normally present in cast steels it cannot lower the FeO content sufficiently to prevent reaction with carbon during freezing.   The theoretical amount of iron oxide in equilibrium with 0.50 per cent manganese at 2900°F. is about 0.2 per cent.[24]

It has been shown that iron saturated with H_2 and N_2 in the molten state will freeze without evolution of gas.   For this reason these gases have a pronounced tendency to form supersaturated solutions as the solubility drops during cooling and freezing of molten steel.   It is difficult to form a bubble of hydrogen or nitrogen in molten steel.   If a bubble is formed from some other source, regardless of how small that bubble may be, H_2 and N_2 will diffuse readily into it and cause it to grow with great rapidity.[22]

Hydrogen can react with FeO according to the reaction

$$H_2 + FeO \rightarrow H_2O + Fe$$
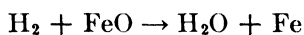
Water vapor, highly insoluble in steel, will readily form a small bubble should this reaction take place.   Whether the reaction does or does not take place depends upon the total and relative amounts of H_2 and FeO present.   For a given quantity of H_2 a certain minimum amount of FeO will react with it.   For larger amounts of H_2, less FeO is needed.   Sims[25] estimates that the best that silicon can do, even when present in a large quantity, is to reduce the FeO content to about 0.01 per cent.   With relatively small contents of H_2, this is low enough to prevent gas evolution.   However, if larger amounts of H_2 are in solution, then an FeO content lower than 0.01 per cent is necessary.

Aluminum is the strongest of the common deoxidizers.   A slight excess of aluminum will reduce the FeO content to 0.002 per cent.   A greater excess will lower the FeO to less than 0.001 per cent (see Fig. 30). With the FeO at such a low value, a wide margin of safety is gained against possible bubble formation.

The evolution of nitrogen may be prevented by the addition of ferro-titanium to the melt, to form iron nitride.[15]   Titanium reacts with iron nitride much as aluminum does with iron oxide, killing the steel with respect to nitrogen.   Titanium is also effective in reducing the FeO content.   Its position as an effective deoxidizer in this regard is roughly halfway between silicon and aluminum.

Other deoxidizers are vanadium, calcium, and zirconium.   These deoxidizers act similarly to silicon, aluminum, and titanium, except in degree.   Vanadium is probably close to silicon in deoxidizing power. There is some doubt as to the effectiveness of calcium.   It may be no more effective than manganese; at least it is certainly not stronger than silicon.   Zirconium probably is about as effective as aluminum but,

because of its violent reaction when adding it to molten steel, its use
has been somewhat restricted.

POROSITY AND CAVITY FORMATION

Gas cavities in steel castings vary in size from very small pin-point
type of porosity to gross blow holes. The shape of these cavities extends
from spherical to elongated ovals, and the inside surface is usually smooth.
The coloring of the surface face is from bright to a dark-oxide luster,
depending upon whether or not the cavity has access to the outside face
of the casting.

Ordinarily, those who produce steel castings are accustomed roughly
to designate gas cavities of large diameter as blow holes, and the small-
diameter cavities (generally less than $\frac{1}{4}$ in.) as porosity holes or pinholes.
It is generally recognized in the steel casting industry that pinholes
constitute one of the chief defects experienced by foundries that specialize
in the production of small steel castings. While this type of porosity
is not confined to small castings, its commercial significance in the lighter
products is greater than in the heavier castings. The standard of accept-
able appearance is generally higher for small castings than for large ones,
i.e., the approach to perfection in appearance desired by the average
consumer becomes greater as the casting becomes smaller. The average
observer is not likely to be critical about a few pinholes in a massive
structural casting, but he would regard them as undesirable in a small
pressure casting. In the light section castings there is also the possibility
that the influence of a gas cavity is potentially more serious with respect
to the casting serviceability and in reducing the established factors of
safety.

Pinhole Porosity.—Pinholes are small (of a size only to accommodate
the point of a pin), elongated, smooth-walled gas holes that occur imme-
diately under the skin of a casting. The long axis is perpendicular to
the surface.[22] Pinholes may occur under all surfaces but often are more
prevalent at reentrant angles. The layer of steel covering them is usually
thin, so thin that when the layer of scale—formed during heat-treatment
of the casting—is removed, the pinholes are uncovered and appear such
as shown in Fig. 31. There are a number of interesting observations
regarding pinholes. A portion of a heat or even a part of a casting may
exhibit pinhole porosity, while the other part will be free of the defect.
Steels that develop pinholes when cast into certain green sand molds
may be free of pinholes when cast into well-dried molds. Steels well
killed with silicon and aluminum may exhibit pinholes, whereas steel
that has apparently been similarly prepared will exhibit no pinhole
porosity.

Sims and Zapffe[22] have discussed at length the theories that have existed on pinhole formation. In 1882 Muller[26] ascribed the formation of pinholes in steel castings to hydrogen precipitated from the advancing layer of freezing steel. The concensus of opinion with respect to pinhole formation was expressed by Batty[27] as caused by mold gases forcibly injected into the casting during the early stages of freezing. The objection to this theory is the presence of the layer of nonporous metal between the pinholes and the casting surface. Woodward[28] explained the mode of entrance of mold gases through the solidified skin of the casting as

Fig. 31.—Pinhole porosity in a steel casting (actual castings).

afforded by the presence of nonmetallic inclusions with low fusion points, which, remaining fluid longer than the steel, give the gases an entrance into the interior of the casting. Arguments have been given that the gas cavities present in steel castings are formed by CO. Caine[29] concluded that pinholes in steel castings are caused by evolved gases, and that the gas responsible is sometimes H_2 but more often is CO.

The rim holes of rimming steel ingots have the same shape and similar position in relation to the surface as pinholes in castings. Hultgren and Phragmen[23] show conclusively that thin blow holes are caused by CO from the reaction $FeO + C \rightarrow Fe + CO$, although there are always small quantities of H_2 and N_2 in the gas evolved. Chipman,[24] Sims,[25] and Sims and Zapffe[22] show the improbability of the formation of CO in dead-killed steel for castings, while Swinden and Stevenson[30] give ample evidence that H_2 can cause porosity in silicon-killed steels.

The theory of pinhole formation, as presented by Sims and Zapffe,[22] appears to agree with all the known facts. They show that pinholes are started by bubbles of water vapor adjacent to the frozen skin and that the bubbles grow simultaneously with, and in the direction of, the steel crystals in the surface layer of the casting by the diffusion of hydrogen into the bubble from the surrounding solidifying and liquid steel, as illustrated in Fig. 32.

The water-vapor bubbles are formed, according to Sims and Zapffe,[22] as follows:

Almost immediately after the molten steel contacts the cold, moist sand, a thin skin of solid steel forms. At the same time the water in the adjacent sand is changed
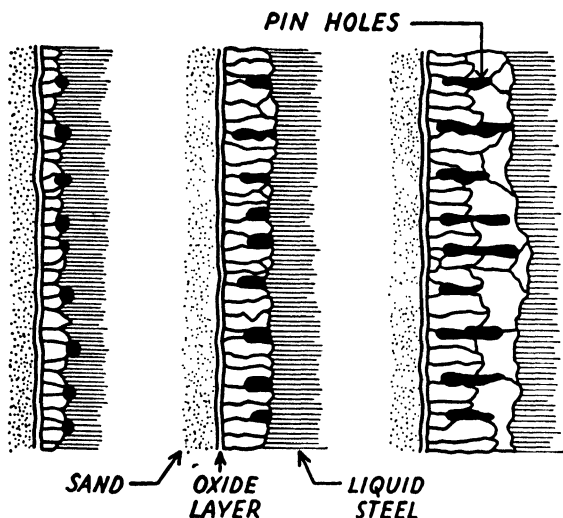
Fig. 32.—The origin and growth of pinholes. (*Sims and Zapffe.*[22])

to steam, with an increase of some 5,000 volumes, and creates an atmosphere highly oxidizing to steel. A portion of this water vapor escapes through the porous sand, but some of it reacts with the hot steel thus:

$$H_2O + Fe \rightarrow 2H + FeO$$
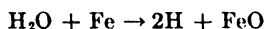
The FeO is formed as a surface layer. The released atomic hydrogen readily dissolves, and the rate of diffusion is so great at temperatures just under the freezing point that it flows through the thin layer of solid steel about as readily as the water vapor penetrates the facing sand. A higher concentration of hydrogen is built up in the liquid steel.

When the hydrogen concentration becomes high, the hydrogen will react with dissolved FeO in the steel to form a small bubble of water vapor on the frozen wall. Immediately the hydrogen from the adjacent liquid steel, as also that still diffusing through the solid layer, will begin diffusing into the bubble and collecting to form molecular hydrogen. Thus, the bubble grows.

Freezing of the columnar crystals of metal is going forward at the same time that the bubble is growing. Bubble growth continues until the hydrogen is depleted from the surrounding metal.

Metal and Foundry Conditions Relative to Pinhole Porosity.—A number of conditions are responsible for the presence of hydrogen leading to pinhole formation:

Effect of Steelmaking on Pinhole Porosity.—An effort should be made to produce steel with a low hydrogen content. The sources of hydrogen in the fluid steel are or include

1. Hydrogen introduced into the charge by wet, rusty, or oily scrap.

2. Water vapor or steam in the air employed for combustion and in the furnace atmosphere.

3. Water vapor formed by combustion of hydrogen in the fuel.

4. Moistened slag-making materials, slaked lime, wet ore, or wet refractory materials.

5. Holding of heats and the arcing on heats in the furnace after the final deoxidizers have been added.

6. Use of wet ferroalloy additions.

Hydrogen absorption from the furnace atmosphere is determined mainly by the humidity of that atmosphere. The humidity varies greatly as the weather changes. All materials added to the bath should be dry, and all water-pipe connections should be regularly inspected to prevent water from leaking into the furnace or refractories.

The only agency for driving hydrogen out of the bath is the so-called "boil." The CO bubbles escaping from the bath act as a carrier of the hydrogen from the bath through the mechanism of diffusion of the hydrogen into the CO bubble. A long, vigorous boil during the oxidizing period is necessary for low hydrogen content. With ample carbon in the metal, such as 0.20 per cent or more, the boil should be brisk; *i.e.,* the escaping bubbles should almost crowd each other.

Effect of Tapping and Pouring Conditions on Pinhole Porosity.— Volatile matter in the furnace spout, wet furnace spouts, insufficient dryness of ladles, lack of steam vents in ladles, and volatile matter in ladle lining can each add to the increase of hydrogen in the melt.

Old sculls in ladles receiving molten metal, exposure of metal to a number of transfers, inadequate deoxidation by special deoxidizers, ladle additions of alloys, and dipping used skimming rods into metal in the ladles may also be responsible for hydrogen pickup or may cut down the effectiveness of the aluminum deoxidizers.

Effect of the Mold on Porosity.—The initial hydrogen content of a steel will influence its tendency to form pinholes. Such factors as excess moisture content, low permeability of the facing sand, casting contours, excess organic bonding materials, or any condition that will build up a

high pressure of water vapor or hydrogen on the face of the casting will contribute toward a flow of hydrogen into the casting and will therefore determine the prevalence of pinholes. Moisture in the mold is probably the most outstanding source of hydrogen, and the steam generated by the molten metal is the most virulent form of inoculator. Green sands and mold washes have long been recognized as sources of moisture.

Thus a steel that is high in hydrogen content may not produce a casting containing pinhole-porosity defects if the mold conditions are such that hydrogen or water vapor is not formed. If hydrogen is not formed by the mold to react with the oxygen of the metal, or if the aluminum deoxidizer has reduced the available oxygen in the metal to a low content, then pinholes will not be found.

It should be kept in mind, as pointed out by Zapffe and Sims,[15] that

. . . molten steel must always be considered as having some threshold hydrogen content that may or may not be able in itself to cause trouble. If not, it will give the metal a certain degree of sensitivity to further hydrogen pickup, such that a very small absorption in some cases might raise the threshold value above the critical value at which deleterious activities of the gas begin.

The Effect of Aluminum Additions to Prevent Pinhole Porosity.—It has previously been shown that aluminum, added to a silicon-killed steel in quantities sufficient to leave a small residual content, will prevent the formation of pinholes. The FeO content of the steel is lowered to such an extent that even the influx of hydrogen from mold conditions cannot raise the concentration high enough to cause a reaction resulting in water vapor. Aluminum in quantities of 0.05 per cent (1 lb. per ton of steel) is usually sufficient to prevent pinhole formation; however, larger quantities of aluminum are often used to prevent the occurrence of low-ductility properties, as will be explained in greater detail under Inclusions.

Aluminum is used in nearly all cases in the production of acid-electric steel for green sand castings. Troy[32] reports the use of 0.10 to 0.15 per cent aluminum to 10,000 heats and claims adequate protection against pinhole porosity if a vigorous boil is used in the melting procedure.

Blow Holes.—Large cavities in steel castings may be caused by (1) entrapped gases, (2) hydrogen, and (3) nitrogen. Since steel for castings is a dead-killed steel, the carbon-oxygen reaction does not take place. If, from faulty and incomplete deoxidation, it should take place, however, then rimming action is possible. Since rimming steel is not produced for steel castings, it is not considered in this discussion of gassiness in steel castings.

Gas entrapped by mechanical means, such as the entrapment of air or core or mold gases in the casting, creates cavities. The presence of

these cavities is the result of poor planning in the molding of the casting as, for instance, the improper location of gates and risers and core and mold vents. A horn gate can produce such a fountain effect that air or gas may be entrapped. A swirl gate can be responsible for the formation of a continuous metal skin so that entrapped air may not be able to exit through the mold walls, and the expanded air prevents the mold from being filled with metal.

A number of conditions are responsible for the mechanical entrapment of gases in a casting. These conditions are listed in the following paragraphs.

Inadequate Mechanical Vents of Molds and Cores.—Every mold and core contains material that will produce gas when in contact with molten steel. The casting may be so designed and the mold so constructed that the gas formed in the mold cavity might not completely escape, in which case the gas would be entrapped by the rising metal. Occasionally cores contain a large total amount of gas-forming materials created by organic binders. Some cores are so shaped or placed in the mold that satisfactory outlet for the developed gases cannot be provided. The venting of molds and cores cannot be overdone. In many cases vents in cores are not so large as they might easily be made for maximum security against the entrapment of gas. The vents in molds should be rectangularly shaped (see Vents, page 297). Some molds will require much more venting than others, and the shape of the mold cavity and the location of gates and risers will have much to do with the venting requirements. Vertical vents are preferable to side vents in permitting gas to escape quickly.

Low Permeability of Molds.—Molds may have such a low permeability that the gas generated cannot readily escape through the mold walls. In this case the mold vents are overtaxed and may not be sufficient to remove the gas before it is entrapped by the rising metal. No information has been presented on what constitutes a low permeability value. The principal reason for this is the change in permeability of a mold (hot permeability) due to the presence of molten steel.

Low Strength of Facing Sand.—Facing sand that may have low green, dry, or hot strength may crumble, drop, or be eroded to such an extent that it will be carried to a part of the mold where the gas coming off the eroded molding material is trapped. Cavity formation that is the result of inadequate feeding should not be confused with gassiness. When the design of the casting and the nature and position of the risers are such as to prevent complete feeding of a section of a casting, the shrinkage cavities that are produced may have the appearance of gas cavities. Center-line weakness or cavities located at fillet positions have frequently been mistaken for gas cavities. Gas blows have been considered responsible
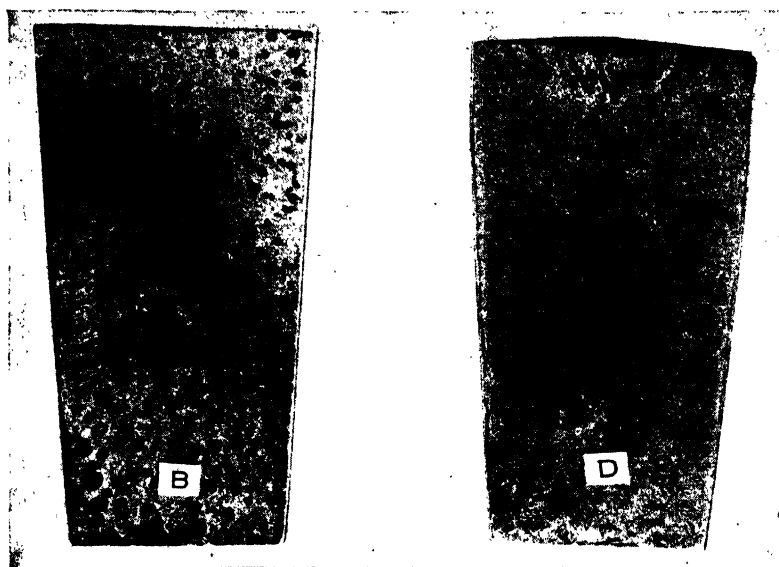
FIG. 33.—Blow holes in a steel casting.

FIG. 34.—Metal rising in molds. (*Zapffe and Sims.*[19])

for single holes at fillets and at reentrant angles; but it has been definitely established that these holes are formed as air passages between the face of the casting and a shrinkage cavity and are the result of atmospheric pressure. The use of atmospheric pressure to assist in feeding purposes is discussed in Chap. VII.

Blow Holes Caused by Hydrogen.—Blow holes are caused by the liberation of hydrogen through the mechanism of reaction and bubble forma-
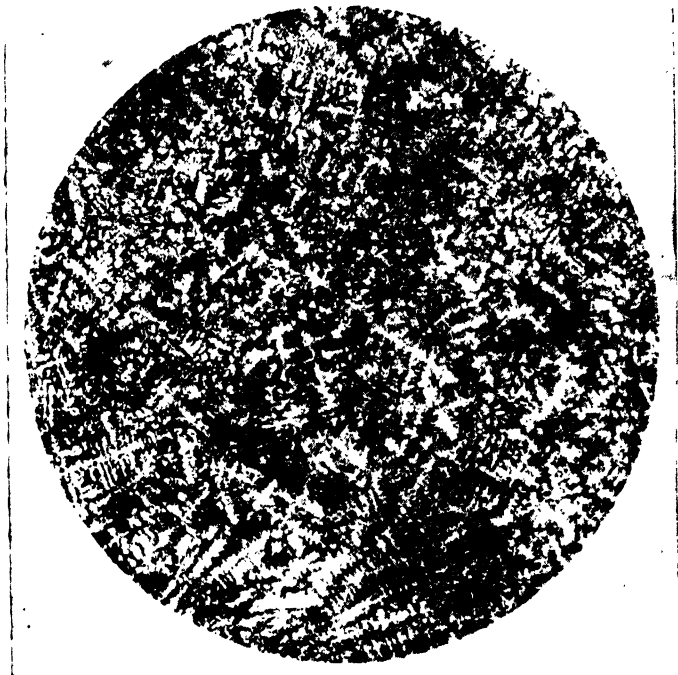
Fig. 35.—Interdendritic voids caused by dendritic segregation by hydrogen absorbed during melting.   X 5.   (*Sims and Zapffe.*[22])

tion. Conditions that are responsible for pinhole porosity may be so magnified in character that the larger gas cavities will be formed. In this case the porosity is not confined to the area near the casting surface but may be general throughout the casting, such as shown in Fig. 33. If the hydrogen content of the metal exceeds the threshold value, blow-hole formation throughout the casting may occur. This can be so excessive that the metal may rise and even flow out of the gates and risers (Figs. 33 and 34). The fact that hydrogen segregates during freezing was demonstrated by Sims and Zapffe.[22] They show that numerous voids may be formed throughout the interdendritic interstices by hydrogen precipitation, as illustrated in Fig. 35.

Moisture condensation on metal inserts, such as chills and chaplets, may be the cause of blow holes in castings. Even though the chills and chaplets are dry when placed in the mold, it is common knowledge that moisture condenses readily on them. Rusty inserts also cause blow holes for similar reasons. It is Sims's[31] opinion that in the great majority of cases the porosity produced by these metal inserts is due to hydrogen in one way or another. It also has been noted that a chill coated with iron oxide may cause no trouble. In the ordinary cast steel there is a sufficient reserve of silicon to prevent any formation of CO gas, unless a very considerable amount of iron oxide is introduced. Furthermore, since steel freezes so quickly when in contact with a chill, iron oxide would be largely confined to a solidified surface layer where it would be inactive. Hydrogen produced from water vapor, however, would not be deterred from passing through this solid layer and entering the molten steel; hence the principal blame for blow holes near metal inserts can be placed on hydrogen.

Blow Holes Caused by Nitrogen.—Blow holes, metal rising, and metal bleeding may be caused by excessive nitrogen content of the metal. The appearance of these defects is similar to that shown for hydrogen in Figs. 33 and 34. Whether or not nucleation plays a role similar to that for hydrogen is not definitely known. The porosity and bleeding caused by nitrogen can be prevented by the addition of titanium as a degassifier.[15]

White Spots (Hydrogen Embrittlement).—The terms "white spots," "snow flakes," "fish eyes," and "flakes" all refer to hydrogen-formed defects resulting from hydrogen embrittlement. An excellent example of white spots in a tensile specimen is shown in Fig. 36.

White spots occasionally are found in cast steels. They generally make their appearance on the tensile-test specimen, although they seldom appear in tensile-test specimens machined from the 1-in. cross-section test coupon, regardless of whether it is separately cast or cast on a casting. On the other hand, white spots are found in the fractures of tensile-test specimens that have been machined from heavy-coupon sections or from castings of large cross sections. White spots conform to the known facts regarding hydrogen embrittlement, *viz.*, a decrease in the ductility of the steel without an appreciable effect on the tensile strength.

The present theory of the hydrogen embrittlement of steel[41] is that normal hydrogen contents of steel can easily supply internal stress condition equivalent to that produced by cold deformation. Then, when the supersaturation of hydrogen in steel exceeds a certain degree, the substructure disjunctions will collect molecular hydrogen under a pressure exceeding the elastic strength and therefore sufficient to force apart the block structure. The stress in the disjunctions is apparently aerostatic and therefore triaxial. Materials so stressed cannot flow; they can only

rupture.   The resulting fracture is different from that found in ductile metals, for under the microscope the white spot is a myriad of tiny, flat, reflecting facets.

It has been stated by Zapffe[41] that if the white spot is given careful examination, the embrittled region will always show some centrally located discontinuity, whether it be a sand or slag particle, a pinhole, a micro-shrinkage cavity, or a small fissure.   The discontinuity may not be found on the fractured face of the white spot, since this face represents only a two-dimensional cross section of the embrittled region.   In cast
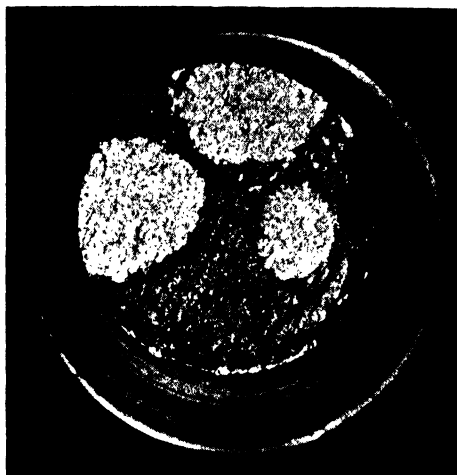
FIG. 36.—White spot in tensile fracture of alloy steel.   X 2.4.   (*Zapffe and Sims.*[41])

steel the embrittled zone is spheroidal and extends below the surface of the fracture.   Its dimensions are determined by the shape of the central discontinuity, and its size by the pressure and quantity of the gas therein.

The fact that the embrittled zone surrounds the discontinuity and that its boundaries are so sharply defined corroborates the theory of Zapffe and Sims[41] as to the formation of these defects.   These researchers present the embrittlement picture as follows:

Hydrogen under high compression in a macroscopic disjunction must also be under high compression in all surrounding disjunctions, regardless of their size, because the gas under those pressures is free to move through the steel: (1) atomically through the lattice and (2) molecularly through the adjoining rifts.

Figure 37 therefore has been drawn to show schematically the probable nature of these defects.   The aerostatic pressure within the macroscopic cavity must be a maximum, and on the surface of the specimen, of course, the pressure must be a minimum.   There must then be some midway point where the pressure is just equal to the elastic strength of the metal, and that point must be the outer edge of the (white spot) embrittled area.   All metal lying between that equistress point and the occluding locus would then be embrittled, and the metal to the other side would be

transferring hydrogen to the surface under pressures that were progressively less as the surface was approached.

Superimposed on the sketch in Fig. 37 is a curve showing schematically how the gas content probably varies over this range. The sharp drop at the edge of the embrittled zone is based upon the reasoning that the rifts probably enlarge suddenly only when the elastic strength is exceeded and are otherwise much smaller in volume.
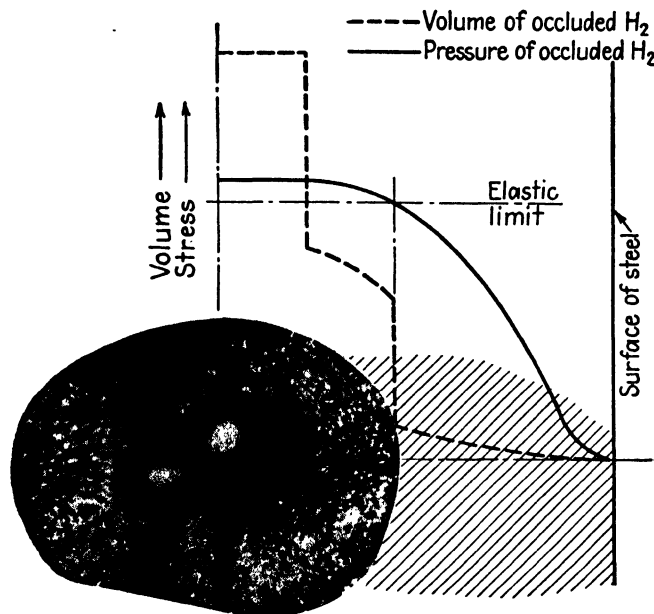
Fig. 37.—Pressure and volume relations of hydrogen in a white spot. (*Zapffe and Sims.*[41])

The spotty test bar also has low-ductility properties. The spots can be eliminated and the ductility values restored to a steel by giving the

TABLE XXXVI.—THE EFFECT OF HYDROGEN EMBRITTLEMENT ON THE DUCTILITY OF CAST STEEL[42]

Section size	Steelmaking	Aged†	White spots	Tensile strength, p.s.i.*	Elongation, per cent	Reduction of area, per cent
4 by 4	Acid induction	No	Yes	96,000	20.2	25.4
		Yes	No	96,000	23.3	41.4
4 by 4	Acid electric	No	Yes	89,000	20.3	35.3
		Yes	No	89,000	26.8	53.6
9 by 9	Basic open-hearth	No	Yes	90,000	18.6	26.8
		Yes	No	90,000	37.7	60.3
9 by 9	Acid electric	No	Yes	84,000	14.8	22.8
		Yes	No	84,000	21.7	42.7

* A normalizing heat treatment given.
† Aging treatment consisted of heating at 400°F. for 16 hr.

steel an aging treatment. This treatment consists of heating castings and test blocks from 400 to 1000°F. for a few hours to 1½ days. As the temperature is lowered, the holding time is extended. The best aging temperature and time for cast steels have not as yet been fixed by experimentation. The final treatment prior to the aging treatment need not necessarily be a fast-cooling treatment to obtain low ductility. A furnace cooling or a tempering treatment may be given to steels showing white spots. The tempering treatments are normally much shorter heating periods than those given to aging treatments. The effect of hydrogen embrittlement and the aging treatment on cast steel in heavy sections can be seen in Table XXXVI.

Nitrogen Embrittlement.—During the early part of the Second World War it was noticed that certain quenched and tempered cast steels showed a very coarse crystalline fracture. Such steels had very little ductility, and it was at first thought that their condition was due to hydrogen embrittlement. This was not definitely proved. On the other hand, it was found that the nitrogen content was high and that the coarse crystalline fracture condition (rock-candy fracture) was found when the residual content of aluminum, after deoxidation, was high. Only a few studies have been made on this subject,[43] and the results obtained are not generally available.

INCLUSIONS

Formation and Type.—The inclusions found in steel castings may be classified into two groups:

1. Those inclusions that owe their origin to the solubility of certain materials in steel and to the chemical reactions of steelmaking.

2. Accidental inclusions, such as entrapped particles of slag, ladle refractories, or molding material.

The accidental type of inclusions is always dangerous, for the size may be quite large, and every care should be taken to avoid their presence. Top-pour ladles should be carefully skimmed and teapot-pour ladles watched for slag streamers during pouring of the steel. Refractory inclusions of ladle brick, nozzles, and runners are usually small, and unless there is excessive wear of these materials, the inclusions resulting from these sources are not extensive. If the erosion is excessive, a change should be made in the refractories used.

Care must always be given to the type and position of gates and molding materials, in order to keep to a minimum the mold erosion and spalling. Information on this subject may be found in Chaps. VII and X. When a steel deoxidized with manganese and silicon freezes, it usually will be found to contain two principal types of inclusion: (1) iron manganese

silicates, and (2) iron manganese sulphides. Often these occur as duplex inclusions consisting of part silicate and part sulphide. The silicates are dark in color, while the sulphides are light gray. These colors vary, since both silicates and sulphides are miscible in the liquid state.

If aluminum is added as a special deoxidizer to a steel deoxidized with manganese and silicon, then upon freezing the steel will be found to contain alumina, silicates, and sulphides. If titanium, zirconium, or vanadium is used as a deoxidizer or an alloy, there is a possibility of the formation of nitride inclusions. Also, in steels that contain chromium, inclusions of chromite ($FeO \cdot Cr_2O_3$) will be found.

Inclusion Types Formed by the Deoxidation with Silicon and Manganese (Globular Type).—The quantity of iron oxide present in the steel before deoxidation controls the quantity of inclusions formed. The addition of silicon and manganese changes the iron oxide into iron manganese silicate, the composition of which varies with the proportions of the constituents.

Manganese combines with FeS to form MnS, although apparently some small amount of FeS is always present. The solubility of manganese sulphide in liquid steel varies with the composition. In low-carbon steel the solubility is influenced by the oxygen content and becomes greater as the oxygen content decreases. For this reason the sulphides form early during freezing and have a globular shape (Fig. 38). The silicates and sulphides are mutually soluble in each other in the liquid state but separate at the time of solidification, even though they often precipitate together.

The glassy silicates (Fig. 38) have a wide range of sizes, distributed at random, and are associated with round or irregular gray sulphide inclusions distributed in an incomplete network. The network is usually poorly defined. When the sulphur content is high enough, the position of the sulphides in the crystal boundaries is clearly shown. Sulphides are found at crystal interstices since during freezing the iron crystallizes first, leaving the metal ever richer in sulphide until only sulphides are left in the crystal interstices. If, because of the presence of active oxygen, which causes lower sulphide solubility, the inclusions start to form early enough, the interstices are wide and the sulphides have considerable freedom of location.

In any section of a casting there is a great variety of sizes of inclusions, but Sims and Lillieqvist[33] have shown that there is a consistent increase in the average and maximum size of the inclusions, with an increase in casting section. The average size of the inclusion in a heat of steel therefore depends upon the rate of solidification. This suggests that the normal inclusion-forming material is soluble in the molten state and precipitates as inclusions during cooling and at the time of solidification.

The globular type of inclusions has been designated by Sims and Dahle[34] as type I inclusions.

Inclusion Types Formed with Deoxidation of Aluminum.—Cast steel deoxidized with aluminum, following the use of manganese and silicon
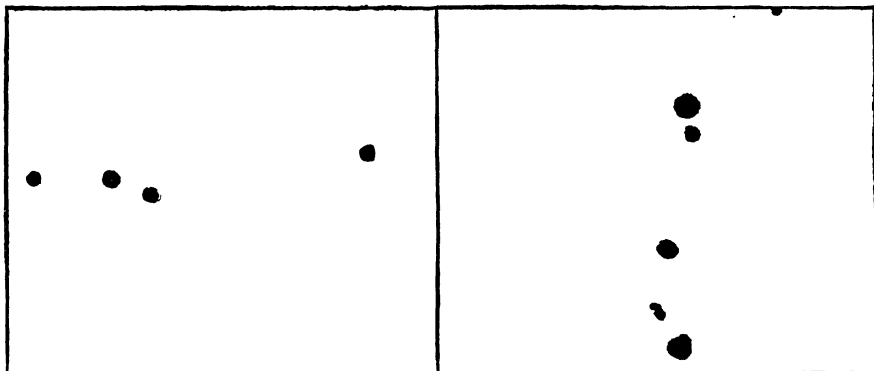
Fig. 38.—Globular silicate and sulphide inclusions. No addition of aluminum. Type I. High ductility. Unetched. × 350. (*Sims and Dahle.*[34])

deoxidizers, will exhibit other types of inclusion. If the aluminum additions are very small, the inclusions will consist of both sulphides and silicates, but instead of assuming a globular shape they will tend to have irregular shapes. This type of inclusion has been designated by Sims and Dahle[34] as type Ia inclusion.

× 185.   × 375.
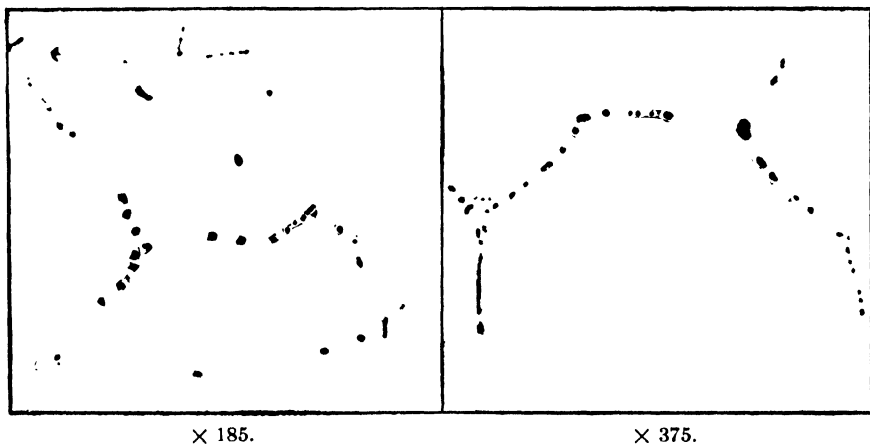
Fig. 39.—Eutectic sulphide inclusions with occasional clusters of alumina. Aluminum added 0.025 per cent. Type II. Low ductility. Unetched. (*Sims and Dahle.*[34])

As the aluminum additions increase, there are a decrease in the quantity of silicate inclusions and a corresponding increase in the fine intergranular sulphide inclusions to produce the eutectic type, or type II[34] inclusions (Fig. 39).

Eutectic Type.—The eutectic type of inclusions is produced with an addition of about 0.025 to 0.05 per cent aluminum to a medium-carbon cast steel. The silicates have completely disappeared, and the sulphides are distributed in a eutectic pattern in the primary grain boundaries. Clusters of alumina inclusions are also to be found.

It has been shown in the previous sections that aluminum is a powerful deoxidizer and will reduce the FeO content to the very low figure of about 0.002 per cent. When the aluminum addition is sufficient to bring about complete deoxidation without appreciable excess aluminum present, the manganese sulphides reach their highest solubility and precipitate as a eutectic with a small portion of iron. The amount of sulphur present will determine the quantity of eutectic. When the sulphur content is that normally found in steel castings (0.03 to 0.05 per cent), only a small quantity of eutectic will be deposited in the primary grain boundaries. Such a eutectic most often will appear as single rows of small, evenly sized and evenly spaced sulphide inclusions, or sometimes it will take the form of discontinuous films. In some cases, such as at the junction of three or more grains, the eutectic appears as a group of sulphide dots or as a lamellar eutectic pattern.[34]

It is claimed by Crafts, Egan, and Forgeng[35] that when the aluminum addition to commercial cast steel is raised to 0.075 per cent, there is a tendency toward compact grouping of the alumina inclusions in clusters or galaxies. They are of the opinion that the galaxies are a ternary eutectic of metal, an oxysulphide compound, and alumina. Sims,[40] on the other hand, points out that all steels to which aluminum has been added, regardless of the amount, contain alumina inclusions that have a strong tendency to form clusters. These oxides are situated within the primary steel crystals, whereas the eutectic sulphides are necessarily in the grain boundaries. The size of the clusters and their distribution are extremely random.

Alumina Type.—When the aluminum additions in commercial cast steels are increased to approximately 0.10 per cent, the inclusions undergo another change. The inclusions consist of randomly distributed alumina and round-to-irregular-shaped sulphides. The sulphides are generally larger and farther apart than those found in type II, but they still persist as a network formation at the grain boundaries. The sulphides appear to contain more than one constituent, and clusters of alumina appear occasionally. These inclusions have been designated as type III by Sims and Dahle[34] (Fig. 40).

If aluminum is added in a quantity so that it is in excess of the oxygen equivalent, a portion of it will form aluminum sulphide (Al_2S_3), which takes the form of a double sulphide with FeS. The effect of the Al_2S_3 is to make the sulphides less soluble, so that they will precipitate earlier in

freezing. Thus the Al_2S_3 tends to prevent the formation of a eutectic and causes the sulphides to form large, irregular-shaped sulphides, which, although they are deposited in crystal interstices, are formed while these interstices are wide. Hence these positions are not restricted to narrow zones.

The Al_2S_3 is a dark, vitreous-appearing inclusion, and the double sulphides are gradations of color from a light gray to black. The lightest portions are recognized as being mainly FeS.[34] If aluminum addition is still further increased, there is a progressive change in the sulphides. The light-colored constituent decreases, and the dark-colored constituent predominates. At the same time, the sulphides get somewhat smaller and

$\times$ 185. $\times$ 375.
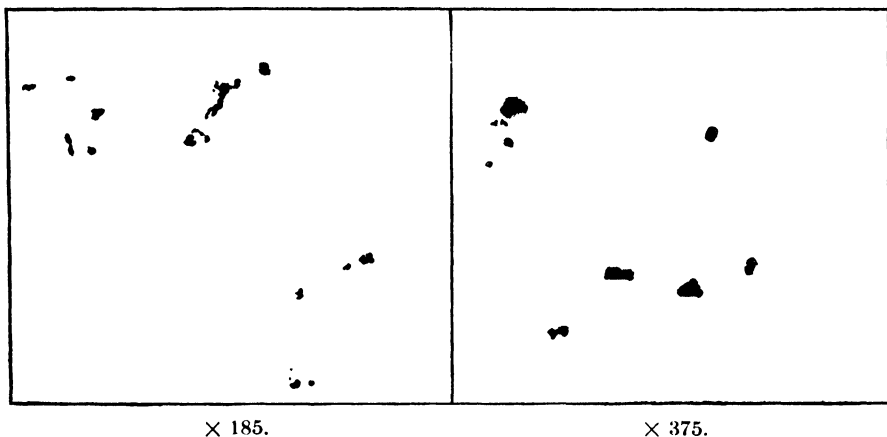
Fig. 40.—Duplex sulphides, large and irregular constituent Al_2S_3. Some alumina present. Type III. Ductility good. Unetched. (*Sims and Dahle.*[34])

more globular in shape. This type is regarded as a modified type III.[34]

Inclusion Types Formed by the Deoxidation with Other Special Deoxidizers.—Any means of depleting the oxygen in steel by the use of zirconium, or titanium, will have an effect on the sulphides similar to that of aluminum. Calcium-bearing alloys added to steel receiving an addition of 0.10 to 0.20 per cent Al, Ti, or Zr produce duplex inclusions and round sulphides that have been classified by Crafts, Egan, and Forgeng as a peritectic type.[35] Both the duplex inclusions and round sulphides tend to form into a network.

It is reported by Caine[35] that inclusions similar in appearance to those formed with calcium additions may be observed after a silicon-aluminum addition. Sims[40] believes that the peritectic type is the double sulphide, $Al_2S_3 \cdot FeS$ inclusion, described under type III and modified type III.

Effect of Inclusions on Properties of Cast Steel.—Steel has not as yet been produced free from nonmetallic inclusions. If the quantity of

inclusions in the range ordinarily found in cast steel is considered, it has been found that they do not have any marked effect on the strength, ductility, and impact values of the steel. It is known that a steel can be "dirty" and still have excellent mechanical properties, provided the inclusions are of the right type.

The consensus of opinion before 1932 was that the presence of aluminum oxide inclusions in cast steel was responsible for the low ductility in aluminum-treated cast steel.[36,37] However, in 1924 Comstock[38] described the chain type of sulphides produced by aluminum additions to cast steel and showed their coincidence with low ductility. This publication was apparently not well known to the industry. In 1932 Sims and Lillieqvist[33] gave considerable substantiating evidence that the type of sulphide inclusions in a steel directly affects the ductility, and that additions of aluminum have a powerful influence on the manner in which the sulphides precipitate. They gave evidence that a steel might have considerable alumina present and still have excellent ductility, providing it also had a favorable type of sulphide inclusion.

It was also believed that low ductility is associated with the use of green-sand molds. Sims and Lillieqvist demonstrated as shown in Table XXXVII that low ductility could be found in dry-sand molds as well as in green-sand molds.

TABLE XXXVII.—MECHANICAL PROPERTIES OF A NORMALIZED MEDIUM-CARBON CAST STEEL WITH AND WITHOUT ALUMINUM ADDITIONS
Sims and Lillieqvist[33]

	Tensile strength, p.s.i. p.s.i.	Yield point, p.s.i. p.s.i.	Elongation in 2 in., per cent	Reduction of area, per cent	Izod impact, ft.-lb.
Dry-sand mold. No Al............	79,000	47,500	31.2	54.6	37.0
Dry-sand mold. 0.05 per cent Al..	79,250	49,250	26.7	40.1	31.0
Green-sand mold. 0.05 per cent Al	79,500	50,000	24.7	35.1	26.5

C, 0.28 per cent; Mn, 0.79 per cent; Si, 0.37 per (, . - , .
per cent.

The eutectic type of sulphide forms a network that causes the steel to rupture after slight plastic flow, with resultant low ductility. Sims and Lillieqvist also pointed out that if the iron oxide content of the steel is high, the inclusions will precipitate at the beginning of solidification and will exist as globules of silicates and sulphides, and the steel will have good ductility. If, however, the iron oxide content of the metal is too low, the solubility of the sulphides is increased to such an extent that they precipitate as a eutectic with the last portion of steel to solidify, and the

ductility of the steel is low. It could be concluded from the results of Sims and Lillieqvist that a heat of steel should be boiled down to a low carbon content (high FeO content in the metal) and then rapidly recarburized and deoxidized just prior to tapping.

Further studies using the above procedure showed that frequently such a practice does not entirely guard against low ductility, and if iron ore were added to the ladle to increase the ductility, the steel casting would exhibit pinholes in green-sand molds.

A complete study of the effect of aluminum on the properties of cast steel was carried out by Sims and Dahle.[34] They found that in the deoxidation of cast steel there is a critical quantity of aluminum that will produce minimum ductility and impact resistance. The critical quantity is the amount that will give complete deoxidation (elimination of FeO) without leaving an appreciable excess of aluminum. The ductility is not affected if smaller amounts of aluminum are added, but such amounts do not deoxidize the steel and safeguard it against pinhole-porosity formation. The ductility of cast steel will be almost as high as when no aluminum is used if the addition of aluminum is large enough to leave an excess.

The sulphur content of cast steel reduces ductility, but the presence of both sulphur and aluminum results in marked lowering of the ductility, as may be seen from Fig. 41. The effect of increasing sulphur is greatest when the aluminum content is that which gives lowest ductility. It can be seen from Fig. 41 that the addition of increasing quantities of aluminum to a steel produces at first a progressive drop in the ductility. Very soon a minimum point is reached where further increments in the aluminum addition cause an upturn in the curves. The recovery of ductility continues to a point only slightly below that of the aluminum-free steel after which further additions of aluminum have little additional effect.

The characteristic shape of the curves is the same for all sulphur contents, but in the high-sulphur steels the loss in ductility is very rapid and of considerable magnitude. The loss of ductility is very slight in low-sulphur steels. In fact, steels that contain as little as 0.02 per cent sulphur show ductilities that are practically unaffected by aluminum additions. The results of Fig. 41 were obtained by making small induction furnace heats.

A number of commercial cast steels were studied by Sims and Dahle,[34] who found that they were unable to detect any consistent trends or relations between ductility and aluminum additions, and melting practice. They were, however, also to relate the inclusion type with the ductility. If no aluminum was added, or if only small quantities of aluminum were used so that no residual aluminum was present but some FeO available, then the inclusions were globular (type I) and the steel

had a high ductility.   The sulphide eutectic type of inclusions (type II) appears with the elimination of FeO and the first traces of residual aluminum.   The ductility of the steel is low with type II inclusions. The large and irregular duplex sulphides of type III appear and occur with type II when the residual aluminum is from 0.01 to 0.03 per cent. The ductility of these steels is not so low as those steels containing type II inclusions.
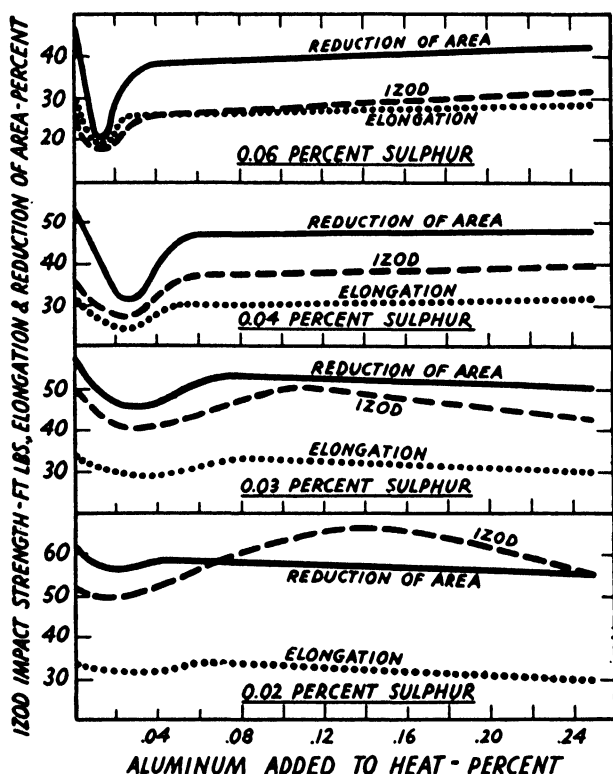
Fig. 41.—Effect of aluminum and sulphur on ductility and impact strength of normalized medium-carbon cast steel.   (*Sims and Dahle.*[34])

A residual-aluminum content of about 0.04 per cent practically assures type III inclusions and good ductility.   Sims and Dahle[34] advised that aluminum additions of 0.05 per cent (1 lb. per ton) were too small an amount to ensure an adequate residual-aluminum content in commercial heats, although sufficient in laboratory heats, and recommended that not less than 0.075 per cent (1½ lb. per ton) of aluminum be used.   Increasing from 1½ to 2 lb. per ton would be somewhat beneficial, whereas larger amounts would neither benefit nor harm.   It is the generally accepted practice in the industry to make an aluminum addition of 2 lb. per ton.[32]

Studies have been carried on by both Sims and Dahle[34] and by Gagnebin[39] to ascertain if other deoxidizers could be used in place of aluminum without forming the sulphide eutectic type of inclusion and obtaining a reduction in the ductility properties. Heats of medium-carbon steel deoxidized with 0.05 per cent zirconium and with 0.05 per cent titanium showed eutectic inclusion and low ductility.

Experiments with calcium and aluminum additions to nickel-mangansese steel showed that simultaneous additions of 0.06 per cent aluminum and more than 0.05 per cent calcium gave the steel excellent ductility properties. Also, excellent ductility results when the calcium alloys are added ahead of the aluminum. The calcium alloys must be forced into the melt, for if added to the surface of the bath, they merely burn away without entering the melt. Other tests by Gagnebin[39] showed that beryllium and zirconium could replace calcium in the calcium-aluminum deoxidation technique. About 0.02 per cent beryllium is equivalent to about 0.1 per cent calcium, while zirconium is used in amounts equivalent to calcium.

For the deoxidizing practice of commercial cast steels, Gagnebin recommends producing steel under thorough-oxidizing conditions and then finishing it with calcium (0.10 to 0.20 per cent as an alloy of calcium manganese silicon) and aluminum (0.05 to 0.10 per cent). The sulphides in the calcium aluminum deoxidized steels are type III and typical of those obtained with recovered ductility when an excess of aluminum is used. The action of the combination of calcium aluminum might be explained as an equivalent in combining power to 0.10 per cent aluminum. However, there also seems to be an additive effect, for calcium is known to form a very stable sulphide that is quite insoluble in steel. Calcium alone apparently is not a strong deoxidizer, but in combination with aluminum it exhibits interesting and useful properties.

Reports from steel casting manufacturing companies indicate that, by using approximately 0.1 per cent calcium and 0.075 to 0.1 per cent aluminum, they are obtaining 100 per cent of the normal ductility in both the normalized and the normalized and tempered conditions. Some foundries have found no advantage in using calcium, and they point out that only aluminum in quantities of 0.1 per cent addition is necessary to maintain excellent ductility and impact properties.

Foundries using the basic open-hearth practice frequently do not use aluminum because of the fact that they do not encounter pinhole porosity in dry-sand or core-type molds. If green-sand molds are used, the steel may be deoxidized with a small quantity of aluminum (0.025 to 0.05 per cent) without encountering the low-ductility problem because of low sulphur content and the oxidized condition of basic open-hearth steel.

Acid open-hearth operators have used for a number of years and are

still using aluminum additions in quantities (0.025 to 0.05 per cent) that, if applied to acid-electric steel of similar sulphur content, would result in low-ductility properties. The exact cause for this is not known, but it is believed that the much longer oxidation period of the acid open-hearth process may be instrumental in changing inclusion solubilities.

The sulphur contents of the steel produced by the basic-electric practice can be maintained at values of 0.02 per cent or below. In this case type II inclusions are not present in sufficient quantities to cause the low-ductility problem.

It has been observed within the steel casting industry that the ductility properties of cast steels fall off considerably with increasing sections, the test specimens in this case being taken from the heavy section. It has further been observed that if these heavier sections (greater than 1 in.) are given an aging treatment (low-temperature draw), the ductility will be greatly improved. It has also been observed that the low-ductility condition is connected with the use of special deoxidizers such as aluminum, titanium, and vanadium, but it appears that aluminum deoxidation is most pronounced in producing low ductility. If, however, these special deoxidizers are not used to obtain higher ductility properties, the likeliness of obtaining porosity defects in the castings is apparently greatly increased.

It is only since the industry has begun to study properties of heavy sections—wherein the test bar is taken from the heavy section—that the problem has assumed major importance. The low-ductility condition is not prominent in the 1-in. test bar, such as the keel-block coupon normally used by the industry to indicate properties of the steel, except that it has occasionally been observed in some green-sand castings. The condition becomes quite evident in sections of 3 to 6 in. It is not known at present whether or not the ductility decreases as the section increases, other than what might normally be attributed to the effect of mass.

It has furthermore been noted that if no special deoxidizers are used and the inclusions are oxide, type I, there is no loss in ductility in heavy sections. Studies have been made on an induction-furnace heat split after deoxidation with ferrosilicon and ferromanganese; one part was cast as deoxidized; the other part was further deoxidized with aluminum and cast. The castings were of similar design. The aluminum-treated portion exhibited low ductility in 4-in. sections, while the nonaluminum-treated steel showed excellent ductility. Steels containing type III and IIIa inclusions in heavy sections always show low ductility that can be improved by aging.

It has been observed that steel made in the acid open-hearth practice may be killed with a special deoxidizer of calcium-manganese-silicon and show no loss of ductility in heavy sections, but it does show low ductility

values if aluminum is used as a deoxidizer. Also, mold conditions may not be a factor, for steels cast in ingot molds show low ductility. An example of the low-ductility effect and the action of deoxidizers in producing low ductility is shown in Table XXXVIII. The fact that an aging treatment is employed to increase the ductility properties would

TABLE XXXVIII.—DUCTILITY OF CAST STEELS IN HEAVY (4- BY 4- BY 10-IN.) SECTIONS[42]

Ladle deoxidation	Aged*	Tensile strength, p.s.i.	Elongation, per cent	Reduction of area, per cent
None............................	No	100,000†	21.7	51.9
	Yes		22.3	51.9
5 lb. per ton FeTi......................	No	95,000	23.0	34.9
	Yes		25.7	48.4
2 lb. per ton Al........................	No	96,000	20.2	25.4
	Yes		23.3	41.4
0.10 per cent V......	No	106,000	17.5	39.6
	Yes		19.0	56.1
1 lb. per ton CaMnSi...................	No	85,000	26.8	57.3
	Yes		27.2	62.6
2 lb. per ton CaMnSi and 2½ lb. per ton Al.	No	86,000	22.8	39.8
	Yes		26.3	50.3

* Aging treatment consisted of heating at 400°F for 16 hr.
† A normalizing heat treatment given.

indicate that hydrogen embrittlement is responsible for the condition (see White Spots, page 144).

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CHAPTER IV

TAPPING AND POURING

Tapping Temperatures.—It is the usual practice to attain the tapping temperature required before the final deoxidation of the steel, since the heat is tapped a few minutes after furnace deoxidation. Metal temperatures are obtained in a number of ways. Some of the methods used do not give true temperatures but record relative values that are correlated with pouring requirements.

Set Test.—The most popular temperature-test method is the set or spoon test. Molten metal is taken from the furnace in a slag-coated test spoon, and the time is recorded in seconds that it takes the steel to film over. The seconds required for any particular heat are correlated with the number of molds to be poured. An actual temperature may be obtained in the spoon test by using an optical pyrometer of the pyro or spot type. A correction based on an emissivity of 0.4 is employed.[4] Any slag coating on the metal must be removed quickly. A difficulty encountered with this method is to find the target quickly and match it almost instantly.

Rod-boil Method.—A cold steel rod is inserted into the bath before the bath is killed. An optical pyrometer is sighted on the metal as it boils up around the rod. The boiling is caused by escape of carbon monoxide from the solution as a result of the local chilling effect.

Furnace-roof Method.—The fuel supply is shut off in the open-hearth furnace or the electrodes are raised in the arc furnace, and an optical temperature reading is made of the furnace roof. Within 15 or 20 sec., the roof temperature will equalize with that of the slag because of radiation. Bath temperature obtained in this way is an average surface indication. If the bath is not homogeneous, readings at individual points, such as are obtained with a thermocouple or spoon, will differ from this temperature. The precision of the method is very good.

Slag-surface Method.—The temperature of the bath is inferred from optical-pyrometer readings on the bright areas of the slag during the time the fuel or power is off. No emissivity correction is applied. These readings are not so reliable as those taken on the furnace roof, since the slag is not a true black body.

Fluidity Test.—A slagged spoon is used to remove molten steel from the furnace. As the spoon comes over the furnace sill, the steel is killed

with aluminum wire. The steel is then poured quickly into a fluidity mold. Correlation charts of metal fluidity vs. temperature, which have been previously prepared by using thermocouple temperature measurements, will show the approximate temperature of the metal. This method has excellent possibilities.

Thermocouple Measurements.—A platinum-rhodium thermocouple enclosed in a refractory tube with a water-cooled head will give accurate temperature readings of the metal bath.[1] A tungsten-molybdenum thermocouple may also be used.[3] Care must be given to the proper preheating of the refractory tube and calibration of the couples. These temperature-measuring methods are very accurate, but they are seldom used in production methods because of the time and care that must be given to the temperature equipment.

Collins-Oseland-tube Method.—The liquid steel is observed with an optical pyrometer through an open tube inserted into the molten steel and is kept clean by a current of air.[2] Readings must be made quickly before the pipe becomes hot enough to bend and melt. The method yields consistent results, and, while it has not been used in the steel casting industry, it has found considerable favor in open-hearth-furnace operations.

Effect of Superheat.—Cast-steel properties are apparently not affected by the degree to which the steel is heated in the molten state. It has been found[6] that superheating has some effect on the primary structure but that in the annealed or heat-treated condition, the mechanical properties show no effect of superheating. Superheated molten steel will, however, have relatively greater gas contents, as was shown in the preceding chapter.

Ladle Temperatures.—As the steel is tapped from the furnace, the temperature is recorded by sighting on the dark portion of the stream by optical pyrometers. The operators of this equipment should station themselves in the same location heat after heat, so that readings will be as nearly comparable as possible.[5] Care must be taken that there is a minimum amount of fumes between the operator and the stream of molten metal.

Since it is possible to measure with a fair degree of accuracy the temperature of molten metal as it leaves the furnace, it is also possible to estimate what the fall in temperature may be between the tapping and the pouring of the first mold. The drop in temperature will depend upon ladle conditions, varying from 50 to 200°F. The magnitude of the cooling is affected by the dimensions of the ladle, the properties of the refractory lining, and the preheating the ladle has received. The time is recorded from the beginning of tapping until the pouring of the first mold. The temperature drop can be estimated for any period of time either by cal-

culation[7] or by the recording and tabulating of previously collected data. A check on the accuracy can be obtained by recording the temperature of the steel as it is poured into the first mold.

Care in Tapping.—In tapping, care should be taken to prevent the slag from entering the ladle along with the metal. It is advisable to do this so that special deoxidizers, such as aluminum, which are added to the ladle, will not lose part of their efficiency by deoxidizing the slag.

A small taphole or a slag trap should be used. One method used is to prepare a 6-in. taphole for each heat. In tapping, the electric furnace is rolled over quickly so that the slag will be above the taphole, as in the open-hearth furnace. If the preparation of slag traps is not feasible, it is suggested that the special deoxidizers be added after the ladle is completely filled by wiring the aluminum in either strip or ingot form to a steel rod and then plunging the deoxidizer to the bottom of the ladle and holding it there until the deoxidizer has melted.

Use of the Furnace as a Holding Furnace or Ladle.—In the triplexing of steel or in continuous pouring, it is often the practice to use the furnace for holding metal. When this is done, small amounts of molten steel, 50 to 200 lb. or more at a time, are withdrawn from the furnace. The temperature of the steel is maintained by occasionally arcing on the metal for short periods of time. Attention should be called to the troubles that may be encountered in such a practice. Deoxidized steel held in a furnace and kept at temperature will usually pick up nitrogen and/or hydrogen, so that it may exhibit pinhole porosity. Tests have shown that steels used in continuous pouring practices are usually higher in nitrogen content than steels tapped in the normal manner.

Ladles.—The ladles used in the steel foundry are of many types and sizes, such as bottom-pour, lip-pour, and teapot type, and shank ladles of lip and teapot-pour types. In some cases there are as many as three transfers of metal, such as tapping from the furnace into a large lip-pour ladle, from there to shanks operated on monorail or by hand, and from there into the mold. The difficulties encountered with ladles in the steel foundry are mostly due to the negligence of proper preparation. A ladle that is improperly lined will erode badly, and pieces of refractory will enter the mold.

The life of a ladle lining will vary considerably with the repair practice used. A firebrick-lined ladle pouring acid steel will give a refractory life of 60 to 75 heats if the ladle is cleaned and repaired with refractory material. Where repair practice involves only chipping off loose slag and applying a light wash, a lining life of from 25 to 35 heats may be expected. With basic operations, the refractory life is much shorter, varying from about 10 to 25 heats.

Ladles are also prepared by ramming refractory mixtures around

forms. One of the best monolithic linings consists of ganister with a fire-clay bond. The life of the monolithic-lined ladles will vary with the mix and type of service, but as a general practice these small ladles are repaired frequently along the metal-refractory contact surface and are relined after about 40 heats. The hand shanks nearly always are prepared with a monolithic lining and are usually relined after being used on each heat.

Defective Castings Resulting from Poor Tapping and Pouring Practices.—Tapping and pouring practices may be such that they are responsible for defective castings. The most prevalent defect found from poor tapping and pouring practices is porosity. The following conditions are known to cause porosity in castings, due primarily to hydrogen pickup by the molten metal:

Insufficient Dryness of the Furnace Spout and Ladle.—The ladle and spout should be thoroughly dry. The practice of daubing a ladle lip or a furnace spout with wet clay or sand should be avoided. The moisture turns into steam and then breaks down into hydrogen and oxygen.

Volatile Matter in the Furnace Spout or Ladle.—The use of molasses or other liquid binders in the lining material may be responsible for hydrogen formation if the lining materials are not completely heated to drive off the gases formed from these materials. Hydrogen is given off when these binders come in contact with molten steel.

Lack of Steam Vents in Ladles.—Ventholes in ladles may be insufficient in number, or they may be plugged with material that prevents the removal of the steam formed when the ladle is filled with steel. If the steam cannot escape through the lining and out through ventholes, it may break down into oxygen and hydrogen, and the hydrogen may be absorbed by the metal.

Old Sculls and Slag in Ladles Receiving Fresh Metal.—Frozen metal in the form of a ladle scull and old slag causes the steel in the ladle to be reoxidized and uses up the special deoxidizers, thereby making the metal less safe and more receptive to the absorption of gases.

Exposure of Tapped Metal to the Atmosphere.—Metal coming in contact with the atmosphere during metal transfer is oxidized. This makes the action of the special deoxidizers less effective. The presence of FeO assists in the formation of porosity as explained previously.

Ladle Additions of Alloys.—All ferroalloys added to the ladle and also to the furnace should be dry, to prevent the formation of steam and its breakdown products of oxygen and hydrogen.

Skimming Rods.—Rods used for skimming ladles often have accumulations of oxidized metal on them. These rods, touching the metal as it enters the molds, are responsible indirectly for the formation of porosity similar to that caused by old sculls and slag.

Poor-grade Refractories, Ladle Sleeves, and Nozzles.—These materials may be responsible for excessive erosion, with the result that the eroded products may lodge in the casting, causing nonmetallic inclusions. In general, sand inclusions from eroded sprues, gates, and molds are largely responsible for this defect, and eroded refractories are usually of a minor nature.

Leaky Stoppers.—Metal spraying from leaky stoppers may deposit metal in the molds in the form of pellets. These pellets may not be reincorporated in the metal because of oxide-film surfaces and give poor casting surfaces.

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CHAPTER V

LIQUID CAST STEEL—LIQUID CONTRACTION AND FLUIDITY

VOLUME CHANGES FROM THE LIQUID STATE
TO ATMOSPHERIC TEMPERATURE

In order to manufacture homogeneous castings that accurately repro-
duce the mold cavity, it is necessary to give considerable thought and
study to the properties of liquid cast steel.

Difficulties experienced in obtaining soundness in steel castings arise,
in part, from the behavior of the liquid steel in the mold. The extent
to which a mold can be accurately reproduced will depend upon the
fluidity characteristics of the molten metal.

In order to provide complete information on the volume changes of
cast steel, a summary of the changes from the liquid state to ordinary
temperature is presented here. A complete discussion of solidification
contraction and contraction in the solid state will be found in Chaps. VI
and VIII.

Steel foundrymen recognize three stages in the cooling of cast steel
from the molten condition to its room-temperature state. These stages
are (1) volume changes that take place in the molten state, (2) the change
in volume on solidification, and (3) volume changes that take place
between the solidification temperature and atmospheric temperature.
These volume changes are known and referred to as (1) liquid contrac-
tion, (2) solidifying contraction, and (3) solid contraction.

Foundrymen are interested in the volume contraction taking place in
the liquid state, since they must take cognizance of this during the plan-
ning of the solidification of a casting. Even greater consideration must
be given to provide against the contraction that occurs during the solidi-
fication of the steel. To disregard this value would result in the produc-
tion of defective castings. Foundrymen are also interested in the change
that occurs in the exterior of the casting as it cools from its solidifying
temperature to room temperature, inasmuch as the rigid form of the mold
hinders contraction and results in the formation of stresses within the
casting, which may develop to such magnitude as to cause casting failure.
The volume changes during these three stages of cooling are shown in
Fig. 42. The effect of the carbon content upon volume changes is shown
in Fig. 43.

It should be mentioned that in commercial practice it is quite possible

to have liquid, solidifying, and solid contraction occurring simultaneously within a single casting, because the molten steel adjacent to the mold walls is chilled and quickly solidified, with the result that solidification proceeds toward the center of the casting. Therefore, the solid outside envelope will be undergoing contraction in the solid state, whereas the portion of the casting that is solidifying is undergoing contraction as a result of the solidification, and the molten metal in the center of the casting is losing temperature and hence contracting in volume.
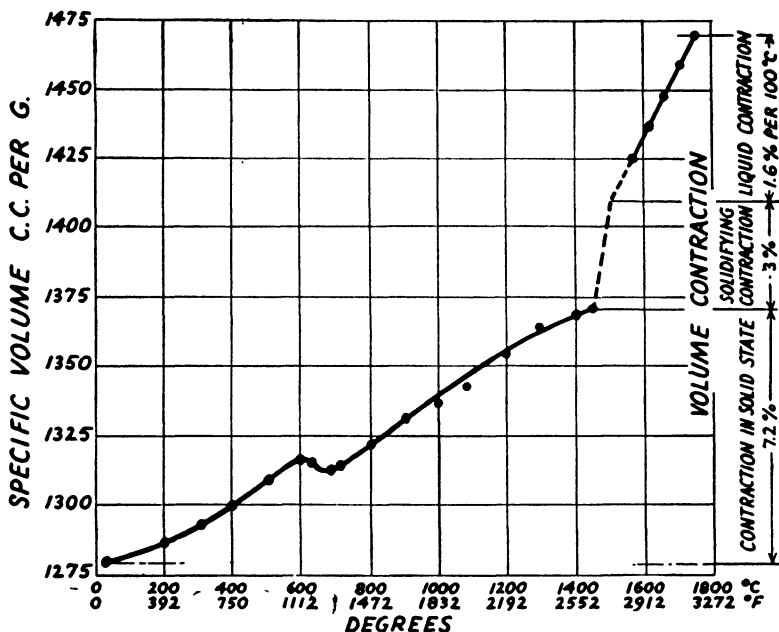
Fig. 42.—Volume change recorded on the cooling of a 0.35 per cent carbon cast steel. (*Briggs and Gezelius.*[1])

Contraction in the Liquid State.—Measurements of volume changes in liquid steel have been few and perhaps somewhat questionable because of technical difficulties encountered when research is carried on at the temperature of molten steel. The most authoritative work on this subject[6] shows the specific volume of 0.30 per cent carbon steel at the melting point of 1494°C. (2721°F.) to be 0.1412 cu. cm. per g., while at 1600°C. (2912°F.) the value is 0.1436 cu. cm. per g. This amounts to a change of specific volume of 0.0022 cu. cm. per g. for each 100°C. (180°F.). Expressed in terms of per cent volume contraction at atmospheric temperature, this is equivalent to 1.6 per cent per 100°C. This value is slightly diminished by an increase of chromium and aluminum and raised by increased amounts of carbon, silicon, manganese, and phosphorus. However, for the usual cast carbon steels and low-alloy cast

steels, the liquid contraction may be considered to be about 1.50 to 1.75 per cent of its volume at atmospheric temperature per 100°C. This value is shown (Fig. 42) in conjunction with the contraction of solidification and the solid contraction.
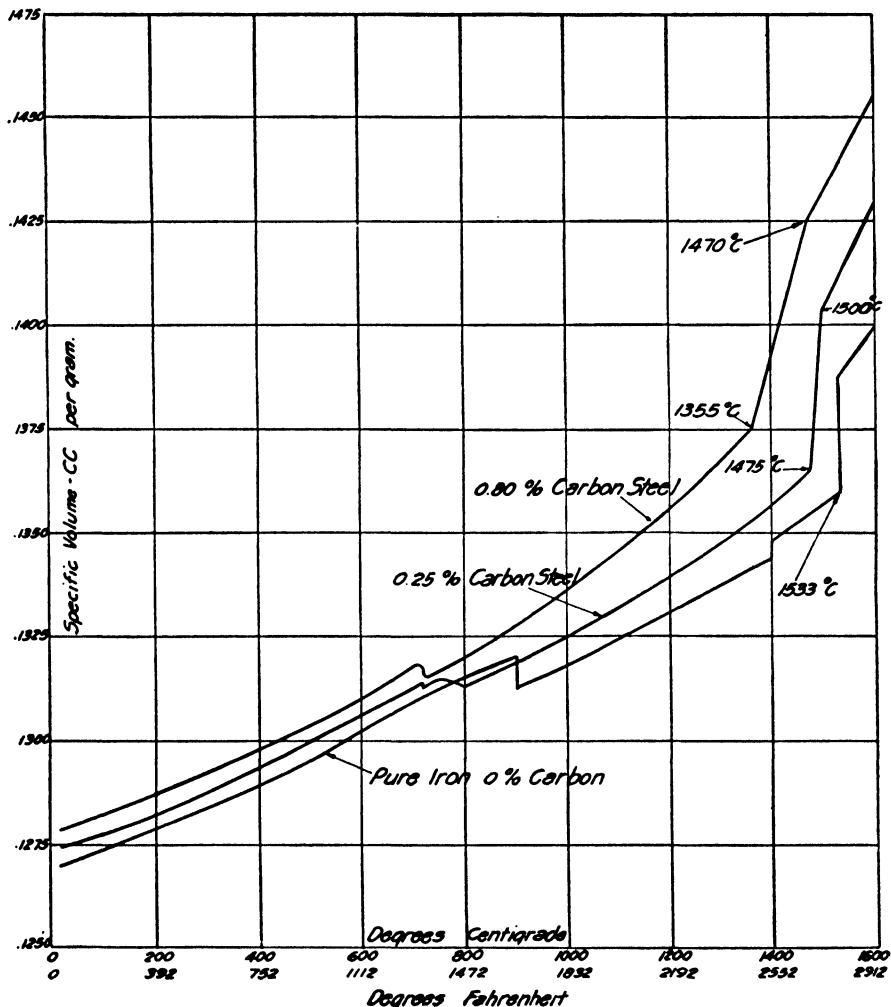
FIG. 43.—Volume changes in iron-carbon alloys. (*Driesen*[2]) and (*Benedicks, Ericsson, and Ericson.*[3])

Contraction at Solidification.—Steel contracts upon solidifying. The various volume-contraction figures that are available show an average of 0.0039 cu. cm. per g. for a 0.35 per cent carbon steel, which is equivalent to 2.85 per cent of the specific volume of the solid steel at the freezing point, or 3.05 per cent of the specific volume at room temperature.

No accurate measurements are available of the effect on the solidifying contraction of such differences as may be encountered by variations in alloy contents in steels.   Experimental evidence shows that pure iron exhibits a solidifying contraction of about 2.2 per cent.   This value increases to approximately 4 per cent for a 1.00 per cent carbon steel. The volume changes brought about by the solidifying of pure iron and carbon steel are illustrated in Figs. 42 and 43.

TABLE XXXIX.—LINEAR CONTRACTION IN FREELY CONTRACTING CAST-STEEL BARS
Briggs and Gezelius[1]

Carbon,* per cent	Contraction above critical range, per cent	Expansion within critical range, per cent	Contraction below critical range, per cent	Total contraction, per cent
0.08	1.42	0.11	1.16	2.47
0.14	1.51	0.11	1.06	2.46
0.35	1.47	0.11	1.04	2.40
0.45	1.39	0.11	1.07	2.35
0.55	1.35	0.09	1.05	2.31
0.90	1.21	0.01	0.98	2.18

* Manganese, 0.55 to 0.80 per cent; silicon, 0.25 to 0.40 per cent.

Contraction in the Solid State.—Measurements of the total contraction of cast steel in the solid state have been recorded lineally on carbon and alloy steel castings.   The solid contraction decreases with an increase of the carbon content of the steel (Table XXXIX).   The contraction

TABLE XL.—LINEAR CONTRACTION IN FREELY CONTRACTING ALLOY-CAST-STEEL BARS
Briggs and Gezelius[1]

No.	Type of steel and per cent of alloy*	Contraction before critical range, per cent	Expansion within critical range, per cent	Contraction below critical range, per cent	Total contraction, per cent
1	Carbon (0.35)	1.47	0.10	1.03	2.40
2	Manganese (1.32)	1.74	0.26	0.88	2.38
3	Nickel (3.00)	1.72	0.20	0.87	2.40
4	Chromium (1.03)	1.46	0.14	1.02	2.34
5	Copper (1.39)	1.49	0.14	1.00	2.35
6	Molybdenum (0.39)	1.65	0.04	0.72	2.33
7	Vanadium (0.25)	1.40	0.09	1.02	2.32
8	Mn-Si (1.35)(1.15)	1.54	0.14	0.96	2.35
9	Ni-Mn (1.46)(1.24)	1.70	0.19	0.86	2.37
10	Mn-Mo (1.41)(0.37)	1.68	0.02	0.66	2.32
11	Mn-V (1.41)(0.16)	1.50	0.13	0.96	2.33
12	Ni-Cr (2.88)(0.91)	1.95	0.04	0.36	2.26

* Carbon kept between 0.32 and 0.37 per cent, manganese 0.55 to 0.80 per cent, and silicon 0.25 to 0.40 per cent, unless otherwise specified.

that occurs before the critical point is reached is in general reduced as the carbon content increases, whereas the contraction below the critical range is about the same for all the steels studied.
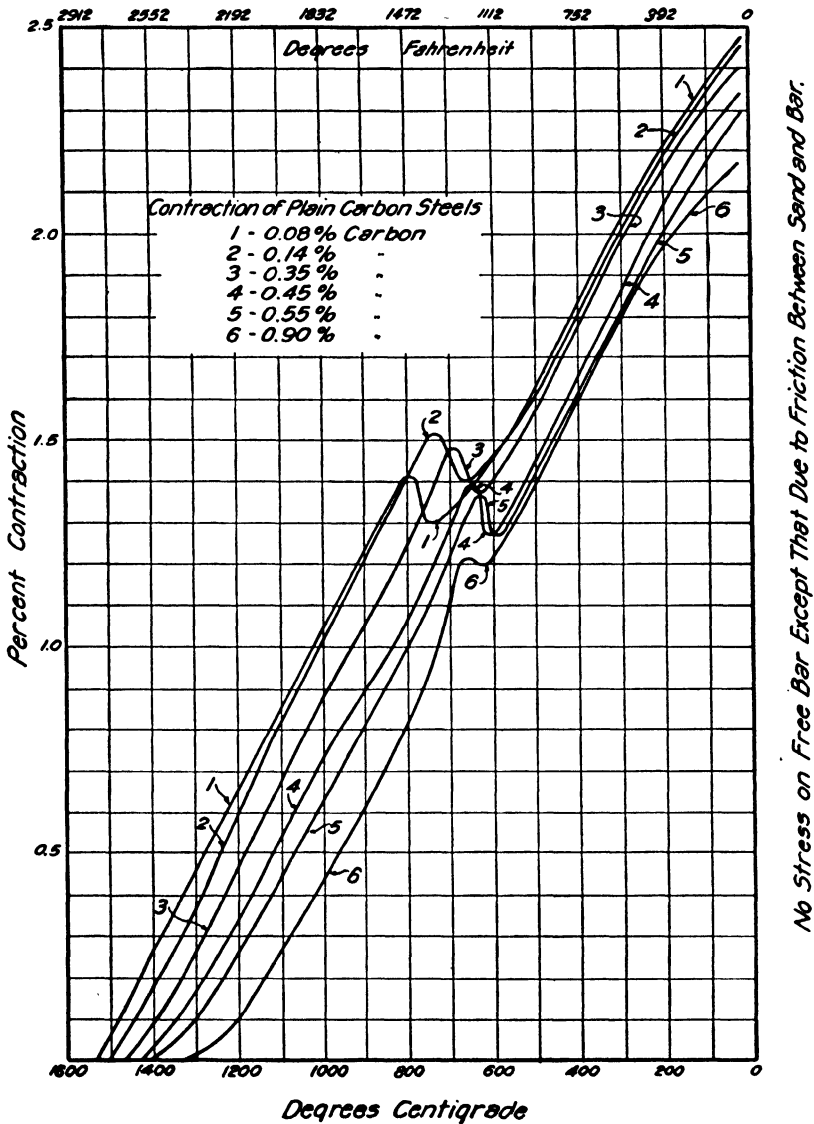
FIG. 44.—Contraction of carbon steel castings.   (*Briggs and Gezelius.*[a])

Alloy additions do not have a marked effect on the total amount of contraction, as may be seen in Table XL, but they do have an effect on the temperature at which the critical range occurs and hence on the amount of contraction that is recorded, both above and below the critical

range.   The type of curves that are obtained is illustrated in Fig. 44.
The linear measurements may be converted into volume changes, as is
shown in Fig. 42.   In this way the specific volume of the cast steel at the
solidification temperature may be obtained.   By converting all linear
values to volume values, it is observed that the solid contraction varies
from about 6.9 to 7.4 per cent.

   Total Contraction.—The total percentage of contraction that takes
place between molten-steel temperatures (*e.g.*, a steel at 1600°C. or
2912°F.) and atmospheric temperature may be ascertained from Fig. 42,
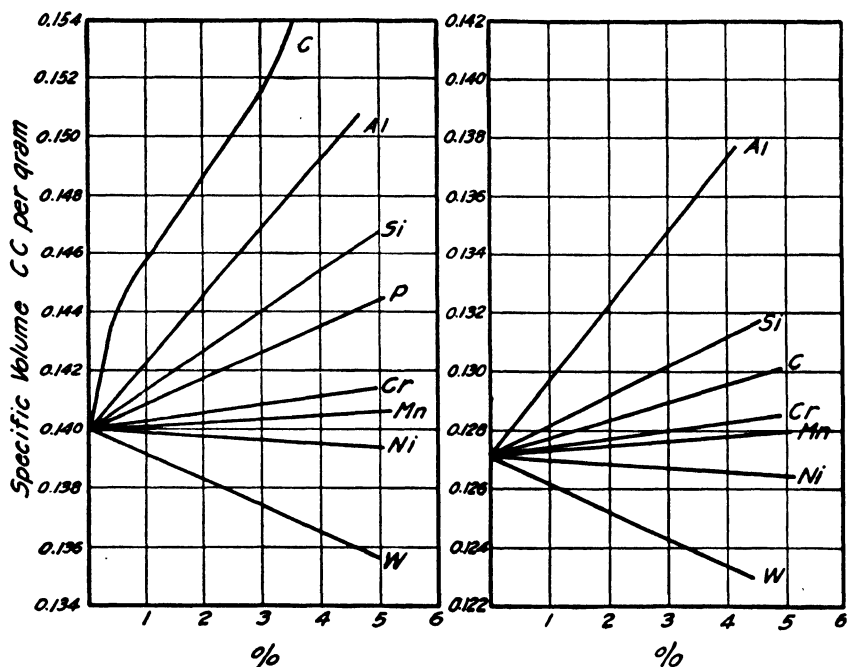
Fig. 45.—Influence of composition on specific volume at 1600°C. (2912°F.) and 20°C.
(68°F.).   (*Benedicks, Ericsson, and Ericson.*[3])

which for a 0.35 per cent carbon cast steel is 11.8 per cent.   This value
changes somewhat as the composition is altered.   A very good estimate
of the influence of composition can be obtained by a comparison of the
specific volumes of steels in the molten state at 1600°C. (2912°F.) and at
atmospheric temperature (Fig. 45).   From these diagrams it is clear that
with the exception of carbon, the additions of other elements influence the
specific volume of the melt at 1600°C. (2912°F.) to about the same extent
that they influence the solid steel at atmospheric temperature.

   From a comparison of these it is evident that the total volumetric
contraction for iron is approximately 10.3 per cent.   The charts also
show that the volumetric contraction of the carbon-free iron alloys is

between 10.0 and 10.5 per cent, a value approximately the same as that recorded for iron. No marked change in the total contraction can therefore be expected from the addition of alloy elements. Of course, the total contraction increases with the carbon content, and reference to Fig. 45 reveals the fact that the total contraction for 0.06 per cent carbon steel is 10.5 per cent, while in the case of 1.0 per cent of carbon the total volume contraction is 14 per cent. Foundry measurements on cast steel, as produced in spherical molds, are in fair agreement with the above figure.[5]

Carbon Per Cent	Total Contraction Percentage of Volume
0.10	10.4
0.45	11.3
0.70	12.1

Defects and Liquid Contraction.—Improper allowance for liquid contraction may increase the size of defects in a section. Steel generally is poured into a mold at a superheat of about 100°F. above the liquidus of the steel, or the temperature at which the steel begins to freeze. If the mold were filled quickly with this superheated steel and no further additions of molten steel made, then at the time the casting solidified there would be a cavity toward the center of the casting, the size of which would be approximately 4 per cent of the volume of the casting. Three per cent of the volume of this cavity would have resulted because of the contraction of the steel at the moment it solidified, and the other 1 per cent addition to the cavity size would have resulted from contraction in volume of liquid steel cooling from the superheat temperature to the liquidus temperature.

It is only when steel castings are very poorly fed that it is possible for liquid contraction to increase the size of cavities formed by contraction. Castings that solidify in such a manner that a reservoir of molten metal can compensate for metal deficiency due to contraction do not contain shrinkage cavities.

Shrinkage cavities are formed by solidification contraction. Liquid contraction may be responsible for increasing the size of these cavities, although it is quite possible that there would be sufficient metal available from some outside source to compensate for that lost by liquid contraction. In this case the cavity would be due to insufficient liquid fed metal to compensate only for the solidification contraction.

THE FLUIDITY OF CAST STEEL

The term "fluidity" refers to the property of a metal that allows it to flow freely and evenly into a mold and fill the mold before such freezing occurs that would offer an obstruction to its further flow. It is a property sometimes referred to as "castability," "runnability," "flowing power,"

or "fluid life." "Fluidity," the term frequently preferred, has a disadvantage in that in the past it connoted a specific physical meaning, *viz.*, the reciprocal of the coefficient of viscosity. However, in recent literature the term "coefficient of liquidity" is preferred for this constant, thus allowing the broader meaning to be covered by "fluidity."

Fluidity in its broader aspects, it must be remembered, should by no means be identified with viscosity. The viscosity of a fluid, in the physical sense, is concerned with movements in the interior of the liquid, and thus it is a definite physical constant. A true viscosity test, which is rather simple and accurate at relatively low temperatures, is exceedingly difficult to perform at the higher temperatures required by molten steel. Foundrymen have therefore found it necessary to devise a more simple method for the study of fluidity. In general, this test consists of pouring molten metal into sand molds that contain a cavity in the form of a spiral. This test is concerned primarily with the fluid life of the metal, and measurements refer to both the metal and the mold.[6]

The fluidity of cast steel is dependent upon two major factors: (1) properties of the metal, and (2) variables of testing. The factors relating to the metal are those of viscosity, surface tension, surface films, gas content, and suspended inclusions. The second factor includes the form of the test piece, the casting head, properties of the mold, rate of pouring, and the degree of superheat.

Properties of the Metal—Viscosity.—A quite general belief is that molten metals differ in viscosity, and that metals that flow poorly have high viscosity. In reality, the viscosity of metals is very low, the kinematic viscosity being less than that of water in all cases investigated.[7] The kinematic viscosity (or viscosity divided by density) is considerably lower than that of water. Since viscosity changes only slightly with change of temperature, it is evident that the viscosity of metals plays only a minor part in their casting behavior, and that the effects attributed to viscosity are mainly due to other causes.

Surface Tension.—When molten metals are poured into a sand mold, the mold surface is not wetted.[8,9] Furthermore, the flow of steel in a narrow channel is impeded by the surface tension of the molten steel. This effect becomes less marked as the diameter of the passage is increased. Surface-tension values[10] of some of the common metals at temperatures slightly above their melting points show that (1) the surface tension of metals increases as the melting point increases, and (2) steel has a very high value of surface tension from 1800 to 2,000 dynes per cm., as compared with that of water, which is only 81 at 20°C.

A high value of surface tension has the effect of increasing the pressure required to cause the metal to enter and flow along a narrow channel; but the direct effect of the true surface tension of many metals (steel prob-

ably included) is almost overshadowed by the effect of surface films, the influence of which is equivalent to that of a greatly increased surface tension.

Surface Oxide Films.—It has been found that surface oxide films affect the flowing power of metals with the apparent surface tension of some metals increased three times. There is some evidence[11] that strongly suggests the existence of films in an overoxidized cast steel killed with aluminum. Such a surface film would reduce the fluidity.

Fig. 46.—Fluidity spiral. (*Taylor, Rominski, and Briggs.*[12])

A chromium-rich oxide skin on steels is also believed to have a marked effect on fluidity. It is difficult to evaluate the effect of oxide films, especially as these may be introduced by the mold.

Gas Content.—One difficulty in applying data relating to the effect of surface tension in the case of steel flowing in a mold channel arises from the fact that the steel is often not in actual contact with the mold but is separated from it by a thin layer of gas, either issuing from the metal or arising from the material of the mold. These gases increase the resistance to flow.

Suspended Inclusions.—Solid matter in suspension causes a rapid rise in the apparent viscosity of a liquid, since suspended particles may

hinder flow. A liquid in which more than a small portion of particles is suspended will have a higher viscosity value than the same liquid free from solid particles. It has been said that dissolved oxides affect fluidity. There is no evidence to suggest that oxide in solution would in itself improve the viscosity of cast steel. It may, however, alter the surface tension or produce some effect on surface oxide films.

Variables of Testing—Form of the Test Piece.—Various types of test form have been used to measure the fluidity of cast steel. It is believed, however, that the spiral test is the most sensitive.[12] The spiral mold developed at the U.S. Bureau of Standards[13] has been accepted in the United States as the standard test for cast irons and nonferrous
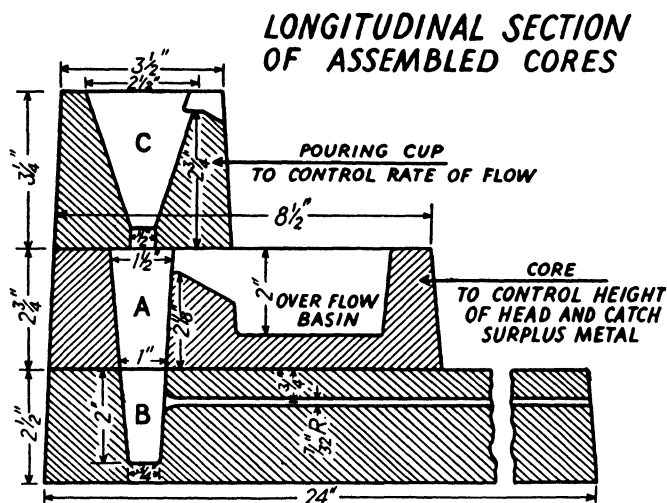
Fig. 47.—Rod-type fluidity spiral. (*Kron and Lorig.*[19])

metals. This test piece has been modified and extended for use on cast steels.[12,14,15] It is not a standard test piece since other experimenters,[16,17,18,19] both in the United States and abroad, have used a straight-bar type of round cross section. The latest developments in the two types of test piece are shown in Figs. 46 and 47. A single test spiral with gating arrangement and casting head is illustrated in Fig. 48. Approximately 20 fluidity tests are necessary to establish properly a fluidity-temperature curve of a cast steel (Fig. 49).

Casting Head.—One of the essential parts of the mold is the casting head and the metal reservoir. The height of these above the test channel affects the length of run. A reservoir must be provided to break the fall of the metal from the ladle and to supply an even flow of metal to the down gate of the mold. The early design of the straight-bar type and some of the spiral modifications used in the study of cast steel did not

incorporate a definite height of casting head. This has been provided for in recent designs of both types of test piece.

Properties of the Mold.—Variations in the temperature of the mold and the character of the mold-cavity surface affect the length of the spiral obtained. In some cases the degree of moisture in the sand affects the

Fɪɢ. 48.—Single fluidity spiral.

length of run, and under other conditions it does not. This depends largely upon the composition of the mold. Variations in the grain size of the sand from which the mold is constructed apparently have no effect upon the length of flow, providing the permeability of the mold is not so far reduced that the gas generated cannot readily escape.[12,16]

Fɪɢ. 49.—Spirals from a single heat ready for checking of length.

Mold wash surfaces can affect the length of flow. In some cases it is increased, while in others a decrease is observed. Coated surfaces can affect fluidity by increasing surface tension or by otherwise modifying the surface films. The mold and coatings also have been observed to affect the degree to which the spiral section is completely filled. The faithful reproduction of mold cavities will also depend upon mold condi-

tions and how these conditions affect surface oxide films, the surface tension of the metal, and the gas formed during the running of the spiral. No definite trends as to mold compositions on these properties are at present available.

Rate of Pouring.—The length of the test bar depends also on the speed of running the metal in the spiral.   The speed, of course, depends on the rate at which the mold is poured.   If the pouring rate varies considerably between one test and another, the length will vary in the same manner. Exceptional care must, therefore, be maintained to pour all molds at the same rate.
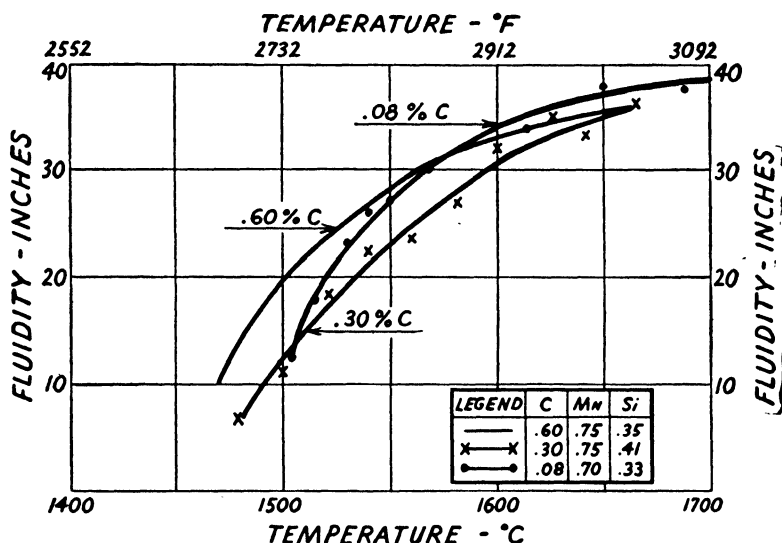
FIG. 50.—Effect of carbon content on the fluidity of steels with average manganese and silicon contents.   (*Taylor, Rominski, and Briggs.*[12])

Degree of Superheat.—As the temperature of molten cast steel is raised, the length of the resulting test bar is increased.   Nearly all reports on the fluidity of cast steel show fluidity curves either strictly proportional to temperature or increasing toward infinity as the temperature increases.   Actually, the spiral length becomes progressively less in the higher thermal ranges.   Investigators have missed this feature because insufficient experimental points were taken and these confined to low temperatures.   Lack of adequate pyrometry was a contributing cause. It is known that in certain cases the degree of superheat alters the composition of the metal and gas content and may increase surface oxide films, for these factors affect the length of spiral obtained.

A very complete set of studies[12] on the fluidity of cast steel shows that carbon has a slight influence on fluidity as it varies from the small amount present in ingot iron to 0.80 per cent.   Maximum fluidity was found for

low-carbon steel, dropping somewhat at 0.30 per cent carbon and restored to an intermediate value at 0.60 per cent, as shown in Fig. 50.

Silicon has a marked effect on fluidity of cast steel, increasing it progressively to roughly 0.60 per cent silicon and maintaining its beneficial influence for additions as high as 1.25 per cent silicon in plain carbon and alloy steels. A marked change toward greater fluidity is found in the medium-carbon steels when silicon is increased from 0.25 to 0.45 per cent (Fig. 51). This is not due presumably to the effect of silicon addition as such, but rather to the state of deoxidation of the steel.
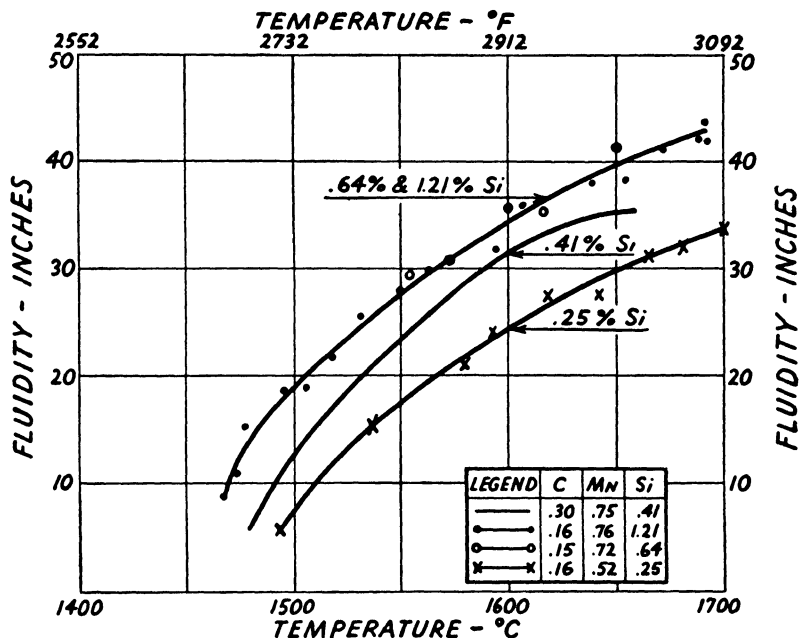
Fig. 51.—The influence of silicon on the fluidity of steels of commercial analyses. (*Taylor, Rominski, and Briggs.*[12])

Silicon in amounts normally present in the commercial cast-steel grades has a much greater effect on fluidity than does manganese. This is due to the fact that silicon is a more powerful deoxidizer than manganese. A manganese-free ingot iron containing 0.30 per cent silicon gives reasonably good fluidity, whereas a silicon-free steel containing a normal percentage of manganese will exhibit very poor fluid characteristics.

An increase of manganese up to 2 per cent has practically no effect upon the fluidity of the alloy over that found in a regular-grade carbon steel, which normally will have a manganese content of about 0.75 per cent. Steels containing manganese greater than approximately 2 per cent, up to 14 per cent, will exhibit improved fluidity at low temperatures (2650 to 2800°F.) as the manganese content increases, whereas at

temperatures of 2900°F. and above, the manganese-rich steels are inferior to plain carbon steels in fluidity.[12]   From the standpoint of fluidity the satisfactory pouring temperature for the 12 to 14 per cent austenitic manganese steels lies between 2650°F. (1454°C.) and 2750°F. (1510°C.).

Copper additions to plain carbon steel increase fluidity progressively as the concentration increases to roughly 4 per cent.   At 6 per cent the length of spiral shows no increase over 4 per cent copper, and 7 per cent copper actually shortens the length of flow.   In percentages normally found in commercial copper cast steels of approximately 1.50 per cent,
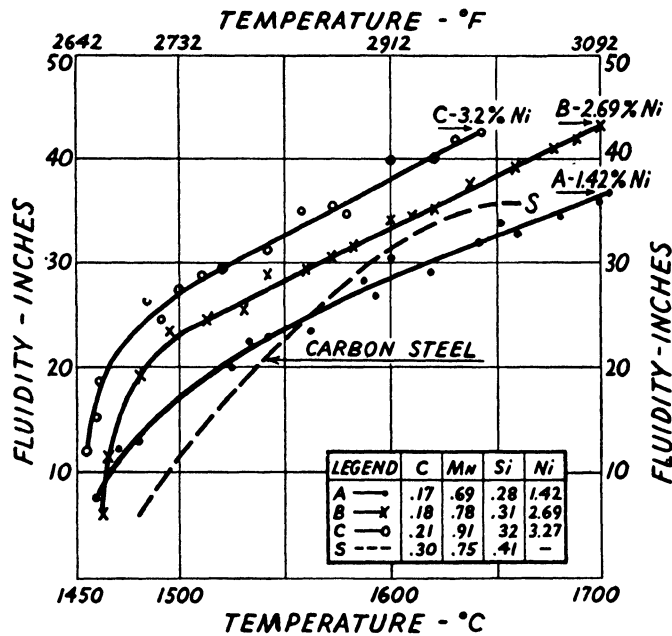
Fig. 52.—The influence of nickel on the fluidity of cast steel.   (*Taylor, Rominski, and Briggs.*[12])

greater run may be obtained at 2850°F. over that recorded for a similar steel without copper.[12]

Nickel improves the fluidity of cast steel progressively by successive additions up to 3.25 per cent.   Higher concentrations up to 5 per cent result in a falling off of fluidity.   Characteristic of this class of steel is a sharp initial increase in fluidity at low temperatures and a steady rise up to approximately 3100°F. beyond the first break at 2700°F. (Fig. 52). Substantial additions of chromium to steel decrease the fluidity as compared with plain carbon steel.   Small additions, less than 1.5 per cent chromium, have no effect on the fluidity characteristics.[12]

Steels containing chromium from 2.8 to 8.8 per cent are nearly identi-

cal in fluidity throughout the temperature range. Except for a small portion at temperatures below 2750°F., the curves for chromium steels are a linear function of the temperature to 3100°F. Molybdenum and vanadium, when added to a normal plain carbon steel in amounts between 0.25 and 1 per cent, cause a reduction in fluidity from that ordinarily obtained if no molybdenum or vanadium were present.[12] Alloy steels of a copper-silicon-manganese composition and a nickel-silicon composition possess very high fluidity. The increased fluidity that these steels possess over a carbon steel is illustrated in Fig. 53.
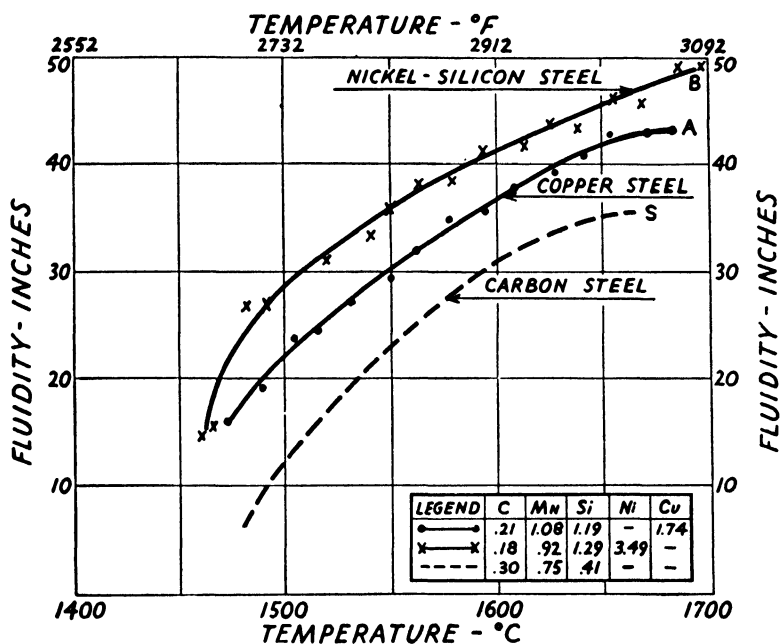
Fig. 53.—Fluidity of special steels alloyed to give maximum fluidity. (*Taylor, Rominski, and Briggs.*[12])

For a weight percentage, aluminum is more effective than silicon in promoting fluidity, especially at the lower temperatures. This is because of the greater deoxidizing power of aluminum. As aluminum is added beyond 0.2 per cent, fluidity is progressively decreased to low values at 1.2 per cent aluminum. This indicates that the amount of aluminum for maximum fluid life is somewhat critical. The threshold value is dependent upon the history and state of oxidation of the bath prior to the addition of the element.

In order to have good fluidity in any cast steel, the steel must be a well-made, thoroughly deoxidized steel. Rimming steels do not have so good fluidity as dead-killed steels. Also, dead-killed steels that are excessively high in deoxidizers such as aluminum or silicon do not have so good

fluidity as steels killed with just a sufficient amount of silicon or aluminum. By adding iron oxide to a steel that has been deoxidized with an excessive amount of aluminum, 1.2 per cent, it is possible to increase markedly the fluidity of the metal.

It is well known by foundrymen that overreduced steel has low fluidity. The fluidity of this steel may be increased by adding scale to the bath just prior to tapping. A certain percentage of iron oxide content is apparently necessary for maximum fluidity, and conditions similar to those show in Fig. 54 are apparently operative. Molten steel that has a high FeO metal content, such as rimming steels, will have a low fluidity; and those having a low FeO content, such as steels made by the
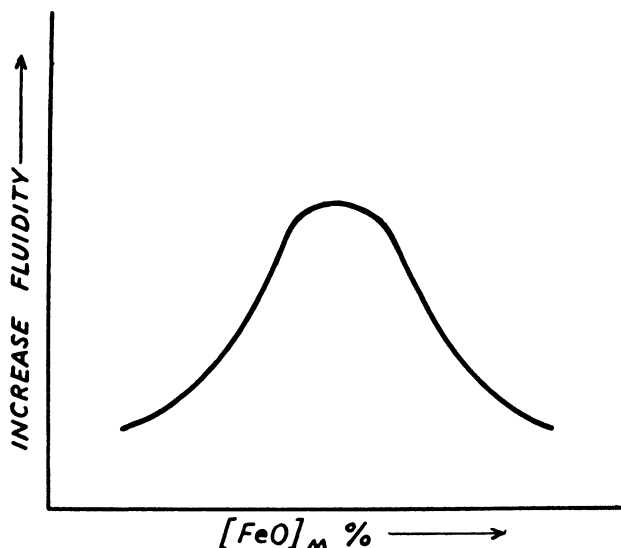
Fig. 54.—General relation of iron oxide content to fluidity.

normal two-slag basic process, will have a low fluidity. The critical amount of FeO is not known. The general form of the curve has not been determined, but it is surmised that it is fairly broad at the top.

Estimates of the fluid character of steels on the basis of visual appearance are likely to be misleading. Some steels that appear dull and sluggish to the eye frequently have better fluid life than others apparently far more lively. It has been difficult to establish a correlation between fluidity tests and actual foundry experience; in fact, little correlation has been established to date, probably owing to the inadequacy of the test methods, for so many factors are involved that only with extreme care is it possible to avoid erratic and meaningless results.

Some foundry studies have been made with both the straight-bar[18,19] and the spiral test-bar[12] types. The test molds are modified so that they

will hold more steel, and reservoir, and gating arrangements are modified so that the pouring of large streams of metal will not introduce large variations such as are encountered when pouring the small laboratory molds. In this way open-hearth heats and electric-furnace heats have been studied on the foundry floor. Steel temperatures have been taken with the optical pyrometer; and in some cases the correlation has not been comparable with conditions obtained on the laboratory-type test mold, where tungsten-molybdenum thermocouples were used to measure molten-steel temperatures.

The limited tests that have been made in foundries are encouraging, for there is a reasonable relationship between temperature and fluidity of the steel. The test-mold designs can undoubtedly be improved for foundry floor measurements. The spiral mold as used in laboratory studies is advocated as a test to be used at the furnace. A routine practice has been set up that permits the tapping of heats based entirely upon fluidity readings. As a large spoonful of steel is removed from the door of the furnace, it is deoxidized with a definite quantity of aluminum and then quickly poured from a standard height into the test mold. Fluidities obtained in this fashion prior to deoxidation of the bath duplicate those obtained after the addition of the final deoxidizers. Thus the melter has a check on the metal fluidity before addition of the deoxidizers. It is believed that this test has considerable merit and that it will find increasing use in the steel foundry.

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CHAPTER VI

THE SOLIDIFICATION OF STEEL CASTINGS

RATE OF SOLIDIFICATION

Nuclei Formation of Crystal Growth.—The solidification of steel begins with the formation of nuclei. Continued solidification consists in the successive formation of nuclei and their growth to the familiar crystals. The number of nuclei that will be formed in a certain quantity of steel is problematic. Examination of sections of cast steel of the same composition and cast at the same time in similar molds will show that the number of crystals does not vary widely.

The distribution of the nuclei is not entirely a random one. When steel is poured into a mold and allowed to solidify, the first nuclei originate in contact with the walls of the mold, since it is here that the temperature first falls low enough for their formation. The numerous nuclei at the mold-metal interface prevent individual nucleus growth in any but one direction, inward, and comparatively few of these grow to such a size as to penetrate far into the molten metal.

The formation and growth of crystals are determined by the temperature distribution in the casting. The temperature distribution may be described at every point, either (1) by the temperature gradient, indicative of the direction of the maximum heat flow, or (2) by the direction at right angles to this, *i.e.*, the isotherm that intersects this point. In any solidifying and cooling casting, there will be a spot or several spots, depending upon the casting design, characterized by a slower cooling than the mass outside. This spot is designated as the thermal center of the cooling casting. The direction of the heat flow is away from the thermal center of the cooling casting.

The first stage in the solidification of a casting is the formation of a thin chill layer, owing to the low temperature of the mold. The chill layer usually possesses a finely grained structure, caused by the rapid cooling. The chill layer, however, may entirely or partially disappear when very hot steel is used or when the mold is heated, reducing the chilling action of the mold. The crystals start to grow from the inner boundary of the chill layer toward the thermal center of the mold. These crystals will necessarily be elongated, for they must grow in the direction of the temperature gradient, and their direction thus indicates the direction in which a casting solidifies (Fig. 55).

183

The mode of casting solidification has been studied by bleeding castings of the remaining molten steel after solidification has begun, at certain time periods after the mold has been poured. After sectioning, the castings show metal skins of varying thicknesses, depending upon the geometry of the mold and the conditions of mold cooling. These skins
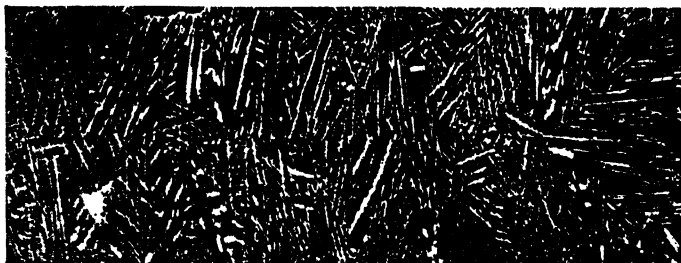
Fig. 55.—Crystalline growth in a steel casting, indicating the manner in which casting solidified. Bottom and sides of casting are mold-metal interfaces. × 7½. (*Courtesy H. S. Jerabek.*)

are similar to heat isotherms, and they depict the flow of heat from the metal during solidification. A diagram of isotherms at an outside corner and at a reentrant angle of a casting is shown in Fig. 56. A casting of the design features of Fig. 56 is shown in Fig. 57 after bleeding at various time intervals. The isotherms that may be found in a spherical casting are shown by the skin formation in the sectioned castings of Fig. 58.
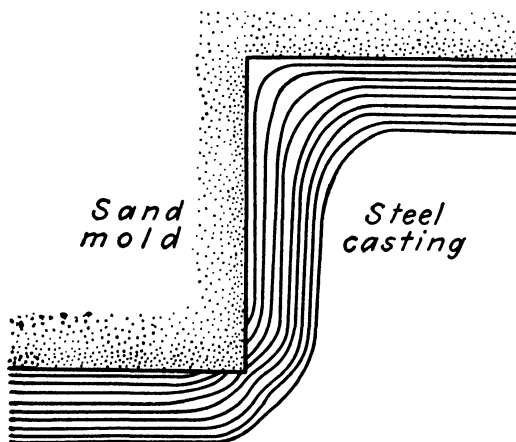
Fig. 56.—Isotherms at an outside corner and at a reentrant angle in a solidifying casting.

The illustrations of Figs. 55 to 58 indicate the nature of the temperature gradients toward the thermal center of a casting and the direction of solidification. In a casting section the direction of solidification has been designated by foundry metallurgists as of two types:

1. Progressive solidification, or that solidification which takes place from the mold-metal interface toward the center of a section.

2. Directional solidification, or the solidification which takes place along the casting member toward an increasing temperature gradient.

Progressive solidification deals with the rate at which solidification proceeds across the casting section or member. It is this phase that will be considered first.

FIG. 57.—Rate of skin formation in joined cylinders. Castings bled after various time intervals. (*Briggs and Gezelius.*[1])

Mold Functions.—The primary function of a mold is to provide a receptacle for the molten steel and also to abstract heat from the molten steel and to dissipate it. The process of abstraction of heat proceeds by two methods that are not entirely independent one of the other. The first is the chilling action that the cold mold has upon the molten steel; the second is the outward passage of heat by conduction through the walls of the mold.

The thermal conductivity of the mold material, its specific heat, and its density affect the rate at which heat can be extracted from the center

of the steel casting. The degree of the initial chilling action of the mold will depend upon the relative specific heats of the steel and the mold materials. For sand molds of the same dimensions, the greater the thermal conductivity, the greater will be the rate of cooling. If the sand molds are made of the same material, then the thickness of the mold walls will govern the rate of cooling. Owing to the poor thermal conductivity of molding sand, the decrease will not be proportional but will tend to a limiting value, beyond which greater thicknesses of sand will have little effect upon the rate of cooling. Conditions that determine solidification, therefore, may be classified as external and internal. The external condi-
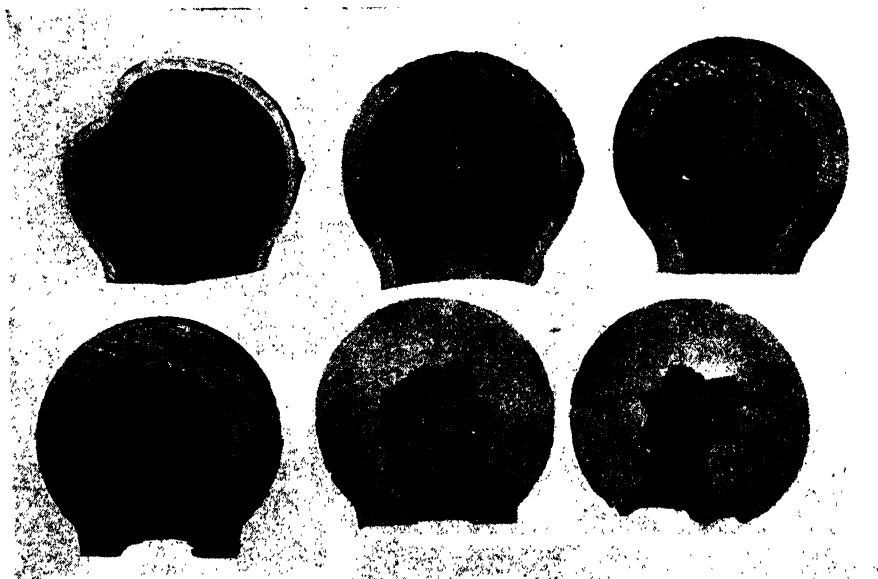
Fig. 58.—Rate of skin formation in spheres. Castings bled after various time intervals.

tions are those that initiate and control the cooling of the steel and are determined by the mold characteristics. The internal conditions govern the cooling effect within the casting and are determined by the physical properties of the steel.

The procedure of solidification of steel, especially of ingot solidification, has been studied by several investigators, notably Desch,[2] Matuschka,[3] Fields,[4] Nelson,[5] Chipman,[6] and others. The Committee on the Heterogeneity of Steel Ingots, of the Iron and Steel Institute,[7] has published a series of reports on the cooling of steel in ingot molds. It is stated that the cooling of steel takes place at first chiefly by a sudden absorption of heat by the mold walls. Subsequently, cooling proceeds by transmission of heat through the mold. These two cooling actions occur simultaneously, but the first starts from a maximum value

and decreases with time, and the second starts from a minimum value and increases with time up to a certain point, when it slowly decreases. The first action (chilling) is capable of freezing at any part of the mold an appreciable mass of steel. The freezing, therefore, can be considered to occur in three successive periods:

1. Almost instantaneous freezing
2. Very rapid freezing
3. Comparatively slow freezing

At the time of rapid solidification of the chill layer, no decrease in temperature in the interior of a large casting can be observed, while the molten metal lying adjacent to the chilling surface will pass rapidly through the range between the superheated temperature and the freezing temperature. The rapid solidification is followed by a slower decrease in the lower superheat temperatures. Because of the low heat conductivity of a sand mold, the velocity of solidification may be so slow that the excess temperature of the molten steel is dissipated before solidification has reached the central portion of the casting.

The chief factors influencing the solidification of steel castings are

1. Solidification factors of steel (*i.e.*, solidification temperature, undercooling effect, latent heat, heat conductivity, etc.)
2. Type, shape, and size of mold
3. Temperature of the steel above its melting point (degree of superheat)
4. Temperature of the mold

These various factors may be considered more fully.

Solidification Values.—Cooling curves for steel[2 and 8] indicate that a state of dynamic heat balance exists in the freezing mass. The term "dynamic heat balance" is used to denote that the latent heat evolved in freezing at any particular moment is just sufficient to balance the heat removed at the surface of the mold.

It has been assumed that, with ordinary pouring conditions, the temperature at the mold-metal interface immediately after pouring is 1100 to 1200°C. (2012 to 2192°F.) in a sand mold and 900 to 1000°C. (1652 to 1832°F.) in a cast-iron ingot mold. In view of the difficulties of experimental determination, the steepness of the temperature gradient between the exterior of the casting and its axis has been largely a matter of conjecture. Calculations have been made[7c,7d, and 9] that quite fully set out the temperature gradients that may be expected in an ingot.

A study of temperature gradients of a spherical steel casting 9 in. in diameter has been made with respect to the temperature gradients in both metal and mold.[1] A temperature of approximately 1300°C. (2372°F.) was taken as the mold-metal-interface temperature during the solidification period of the casting (Fig. 59). A difference of 350°F.

between the center of the sphere and the mold-metal interface was recorded during the first 20 min. of cooling of the casting. Such temperature gradients may be too large, for Chvorinov[10] has established by experimentation and calculation that the surface of a $3\frac{1}{8}$-in.-thick cast plate to be 1450°C. (2642°F.), and that a gradient of only 75°F. exists.

The rate of solidification of steel in sand molds must be determined for castings of simple design in order that data can be reproduced with a fair degree of accuracy. Furthermore, since the type of molding material can possibly change the rate of solidification, a uniform grade of sand
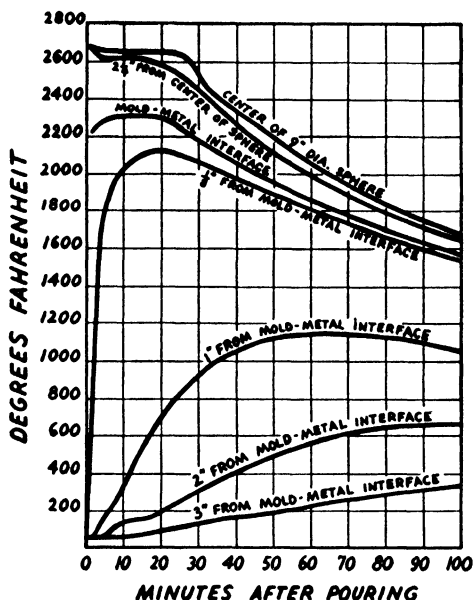
FIG. 59.—Temperature gradients in metal and mold in a 9-in. diameter-sphere casting (*Briggs and Gezelius.*[1])

should be used. Studies carried on in dry molds made of a crude sand synthetically bonded of a permeability of 87 cu. cm. per min. showed the effect of volume and surface area on the rate of solidification in various-size spheres and parallelepipeds.[1] The rates of solidification were obtained on 3-, 4.5-, 6-, and 9-in.-diameter spheres in order that the effect of volume on the rate of skin formation might be noted. These values are presented in Fig. 60. The rate of skin formation increases as the volume of the sphere decreases.

The effect of surface area on rates of solidification was determined by selecting the volume and surface area of the 6-in. sphere as the base and constructing various-sized parallelepipeds having the same volume as that of the 6-in. sphere but with 35, 60, and 110 per cent greater surface area.

The experimental values are illustrated in Fig. 61 and show that the greater the surface area, the more rapid is the rate of skin formation. The differences obtained are small, and, in the case of the parallelepiped,
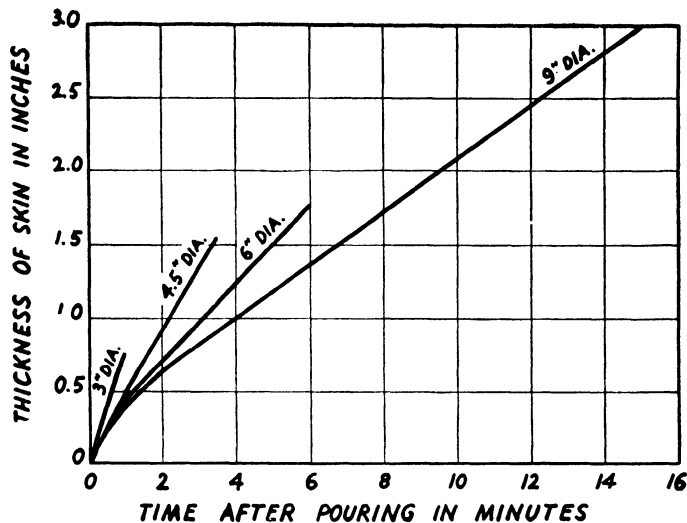
FIG. 60.—The rate of solidification of cast-steel spheres. (*Briggs and Gezelius.*[1])
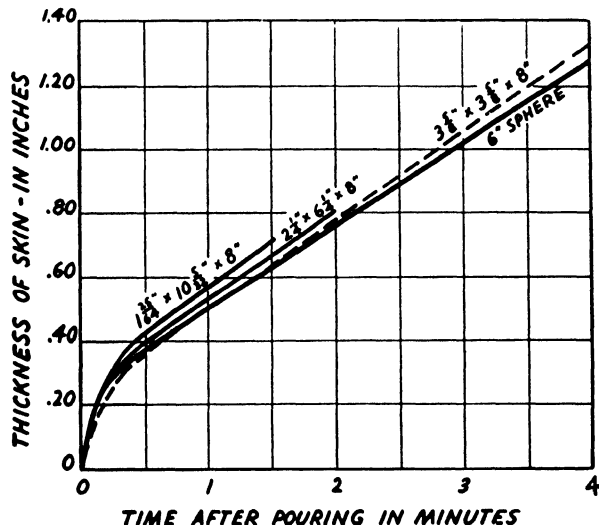
FIG. 61.—The rate of solidification of cast-steel parallelepipeds compared with the solidification rate of a 6-in.-diameter sphere. (*Briggs and Gezelius.*[1])

with 35 per cent greater surface area, the increase in rate is nearly negligible.

It should be noted, as shown by Fig. 62, that even though the linear rate of skin formation varies with the volume of the spheres and also, to a
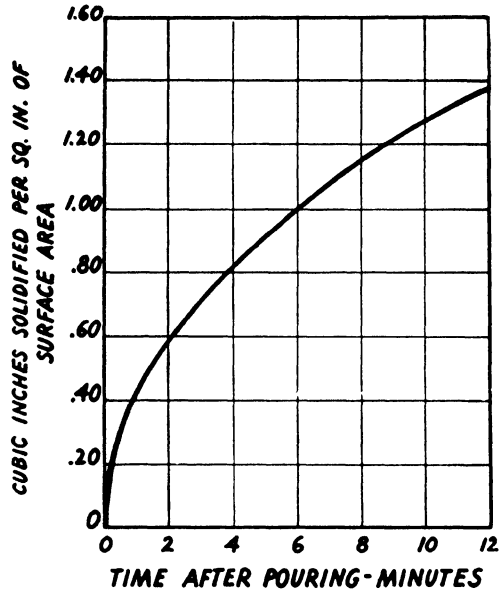
Fig. 62.—Volume of steel solidified per unit of surface area for dry-sand molds. (*Briggs and Gezelius.*[1])
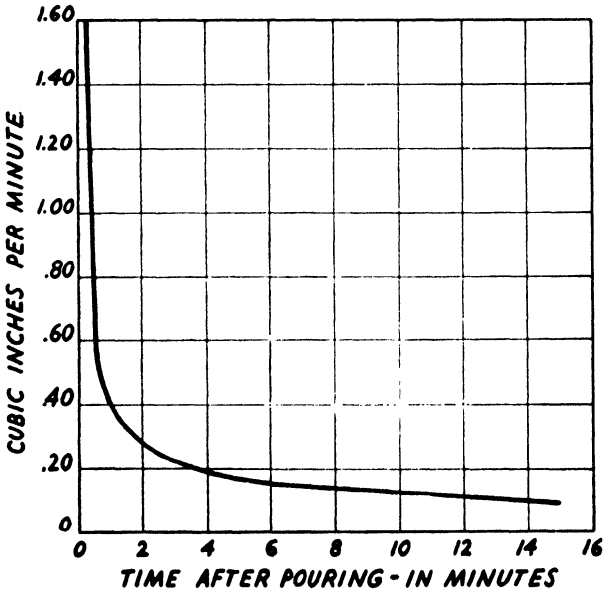
Fig. 63.—Rate of solidification of steel in dry-sand molds based on volume solidified. (*Briggs and Gezelius.*[1])

slight degree, with the surface area of solids with identical volumes, the velocity of solidification is a constant. That is, at any definite time after pouring, the weight of steel solidified or the volume of steel solidified per square inch of mold surface is the same. From this curve it is possible to predict with a fair degree of accuracy the skin thickness of similar shapes at various times after pouring.

It was stated that there are three phases in the solidification of a casting: (1) almost instantaneous freezing, (2) very rapid freezing, and (3) very slow freezing. These transitions are shown in Fig. 63, where the rate of solidification in cubic inch per minute is plotted against time. The initial rate of solidification, due to chilling by the mold, is very high, but it decreases very rapidly to a second phase of solidification that is rapid, which in turn decreases to a very slow rate of freezing.

The segregation of inclusions in steel castings is similar to that found in ingots of comparable sizes. Changes in chemical composition and mechanical properties are detectable if study is made of large sections. There is usually a slight drop in carbon and manganese between the center of a large section and a corner. The strength and ductility of heavy sections are lowest at the center and highest at the surface. The drop in properties will depend upon the composition and upon heat treatment; but a fall-off of 5 to 10 per cent in the properties can be expected in cross sections approximately 8 in. square and larger.

Solidification Calculations.—The solidification of ingots has been studied theoretically,[4,7c,7d,9,11] and the velocity of solidification has been calculated. All investigators have arrived at somewhat the same formula, which tends to show that the velocity of solidification is a parabolic function. The following formulas have been obtained:

Field:

$$D = K \sqrt{t}, \quad \text{with} \quad K = 0.88$$

Lightfoot:

$$D = K \sqrt{t}, \quad \text{with} \quad K = 0.34$$

Schwartz:

$$D = \frac{1}{2} \frac{K}{\sqrt{t}}$$

Here D is the thickness frozen in time t, and K is a constant of solidification.

Nelson[5] attempted to fit the results of the bleeding of ingots to the formula developed by Fields[4] and found that the data corresponded very well in the early stages of solidification, but that as solidification proceeded, his data deviated further from the theoretical curve. Chipman[6] studied the rate of solidification of 18- by 39-in. ingots and cal-

culated from the experimental data that the equation for the curve was

$$D = -0.12 + 0.90 \sqrt{t}$$

An attempt was made to fit the formula $D = K \sqrt{t}$ to the experimental data on steel cast in sand molds. It was found that this equation could be used successfully only on the lower portion of the curve. The data appear to fall upon a hyperbolic rather than a parabolic curve, but a fairly good fit can be obtained using the general equation

$$D = K_1 t^{0.4} + K_2 t$$

where K_1 and K_2 are constants depending upon the shape of the mold, etc. The equations found to represent the spheres are as follows:

Sphere Diameter, In.	Equation
9	$D = 0.34 t^{0.4} + 0.12 t$
6	$D = 0.34 t^{0.4} + 0.17 t$
3	$D = 0.31 t^{0.4} + 0.43 t$

The constant of the first term controls the lower portion of the curve and indicates the chilling action of the sand. This constant is practically the same in each case. The constant of the second term of the equation, which controls the upper, straighter portion of the curve and represents the solidification due to the heat conducted away by the mold, decreases as the volume of the casting increases.
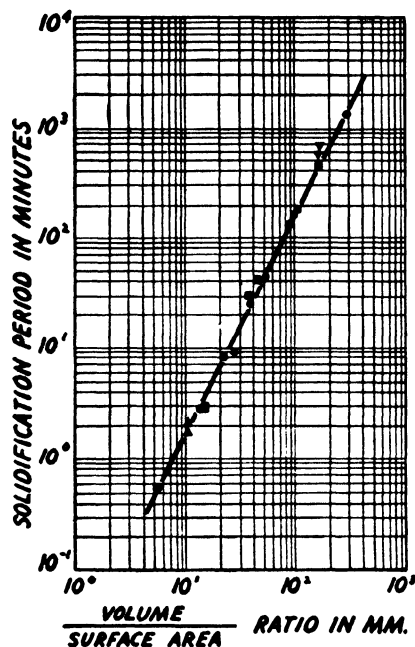
Fig. 64.—Solidification period related to ratio of volume to surface area. (*Chvorinov.*[10])

The problem of the solidification of steel in sand molds was treated slightly differently by Chvorinov,[10] who gives the following equation:

$$t = \frac{R^2}{K^2}$$

where t is the solidification period, *i.e.*, the period from filling the mold to the end of solidification of the last portions of metal, and R is the ratio of volume to surface area. The value of K is calculated to be equal to $K = 2.09$ in. if R is expressed in inches. In Fig. 64 the line represents Chvorinov's equation and is constructed from calculations and experimental data. The solidification period embraces the entire range of

castings as regards both shape and size from a plate 10 mm. (0.39 in.) thick up to a massive casting of 65 tons. Care should be exercised in applying values obtained from the curves of Figs. 62 and 64, since the shape of the casting will influence the results.

Types, Shapes, and Size of Molds.—The types of mold or mold composition do not vary widely, for in most cases they are composed of sand. Some types of mold are made in such a manner that they will collapse after pouring. In such cases relieving blocks, coke, straw rope, and the like are placed in the backing sand. These materials break up the continuity of the sand extending from the mold-metal interface to the
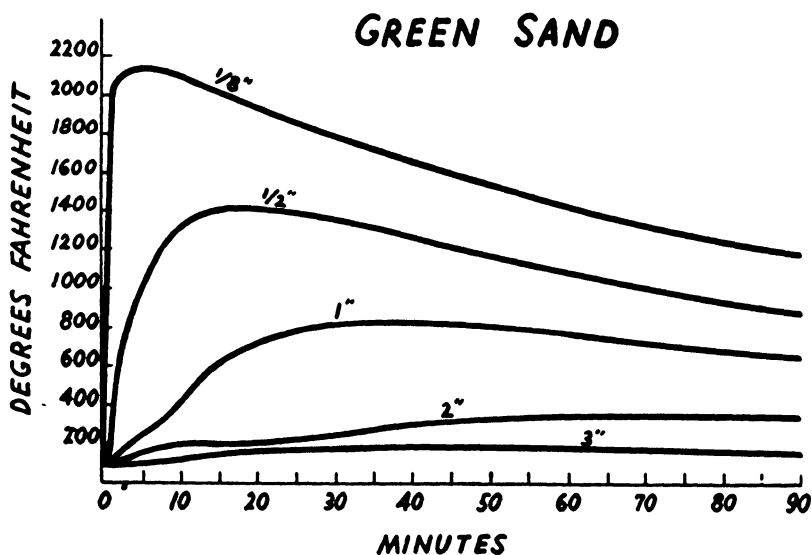
FIG. 65.—Heat-transference rates of a green synthetic-bonded washed silica sand. Permeability 173. (*Briggs and Gezelius.*[1])

exterior of the mold and thus change the termal conductivity of the mold. Whether or not there is a perceptible change in the velocity of solidification or the rate of skin formation has not been studied, but it is thought that such a change would be very small unless these materials were very close to the mold-metal interface and the casting was quite large.

Changes in the solidification velocity may be obtained by the use of various mold materials, but, unlike the ingot mold where the use of copper molds leads to a considerable increase in the velocity of solidification over the usual cast-iron mold, the variations in sand molds result in little more than a perceptible change. Rate-of-skin-formation studies have been made on spheres molded in the same sand with one-half of the mold oven-dried and the other half produced in green sand. The molds were

poured alternately and all bled at the same time interval.   The averaged
results were practically the same.   Thus the rate of solidification is
practically the same in either of the common types of green or dry sand.
This is understandable, since the rate of heat transference of a sand mold
of fair permeability will be about the same, regardless of whether it is
dry or contains 3 to 5 per cent moisture.   A few studies have been
reported by Womochel and Sigerfoos[12] on the effect of variations in
molding sand on the solidification of small steel castings.   Their results
indicated that variations in moisture, compression strength, and mold
hardness had little effect on solidification characteristics of steel.
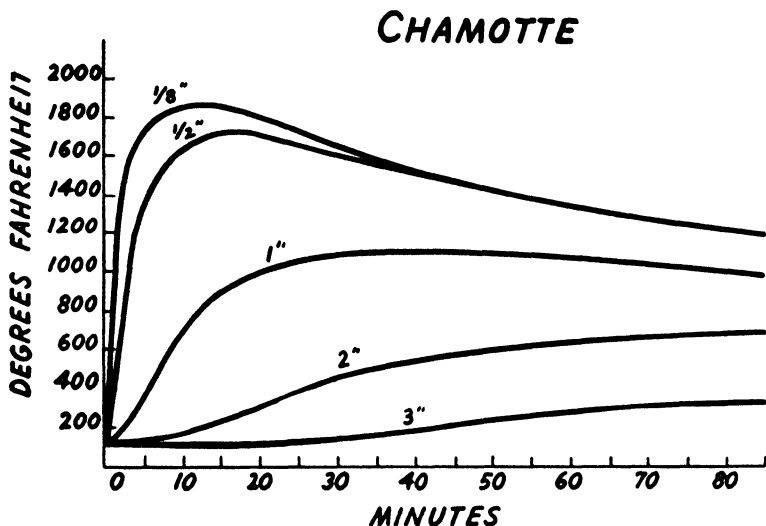
Fig. 66.—Heat-transference rate of chamotte-type sand.   Permeability 2,000 in green state.
(*Briggs and Gezelius.*[1])

The rate of heat transference in sands of different types can vary
greatly as shown in Figs. 65 to 67.   Chamotte, a graded dead-burned fire
clay, has a very high rate of heat transference, while the cement-bonded
sand of Fig. 67 has usually low heat conductivity owing to its low per-
meability.   Sands of these types would change the rate of skin formation.

It seems probable that the rate of skin formation varies directly with
the thermal properties of the mold material.   A formula has been pro-
posed by Chvorinov[10] to evaluate the solidification constant for differ-
ent thermal characteristics of molds:

$$M = 1.158 \frac{T \cdot b^1}{Q \cdot S}$$

where M is the solidification constant, T the solidification temperature

[1] Dimensional units = meters, calories, kilograms, and hours.

of the metal, b the heat-diffusivity coefficient of the mold equal to $\sqrt{K \cdot c \cdot p}$ (where K is the heat conductivity, c the specific heat, and p the volume of mold), Q the heat content of the metal evolved on solidification, and S the specific gravity of the metal at the temperature just after solidification. This formula shows that the amount of metal solidified after a certain time varies directly with the heat-diffusivity coefficient
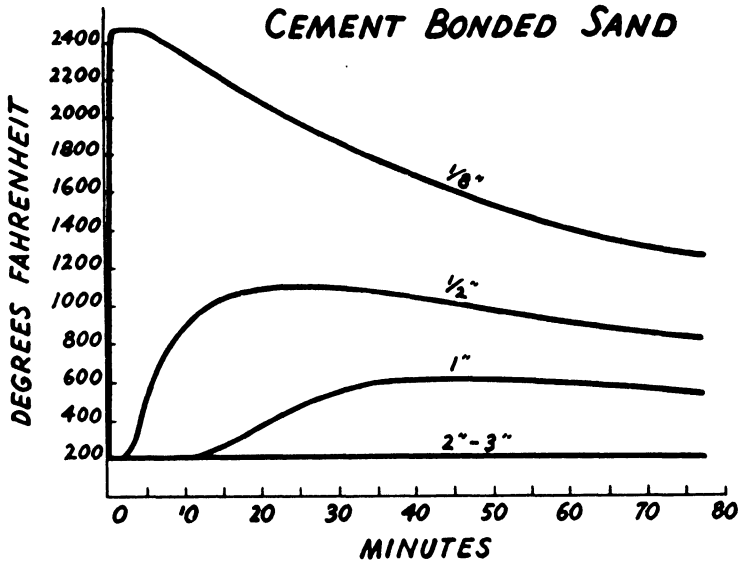
FIG. 67.—Heat-transference rate of cement-bonded sand. Permeability 87. (*Briggs and Gezelius.*[1])

of the mold. Some examples of the heat diffusivity of different materials are shown in Table XLI. The differences that mold material may make

TABLE XLI.—ESTIMATED THERMAL PROPERTIES OF MATERIALS USED AS MOLDS Chvorinov[10]

Material	Heat-diffusivity Coefficient
Steel	170
Mold-sand average	17
Chamotte brick	23
Magnesite brick	45
Silicon-carbide	
Solid	83
Bonded with clay	39

in the solidification constant are illustrated by the example shown in Fig. 68.

The effect that cores have upon the rate of skin formation on that portion of the section touched by the core cannot be definitely defined.

It is believed that cores of a thickness of over four times the thickness of the cast sections that surround them may be considered in the same manner as molds, their cooling capacity being fully adequate as far as the solidification of that section is concerned. Cores of a thickness under the size of the surrounding sections must be considered as exerting a diminishing cooling capacity. With such cores the rate of solidification may be retarded considerably.
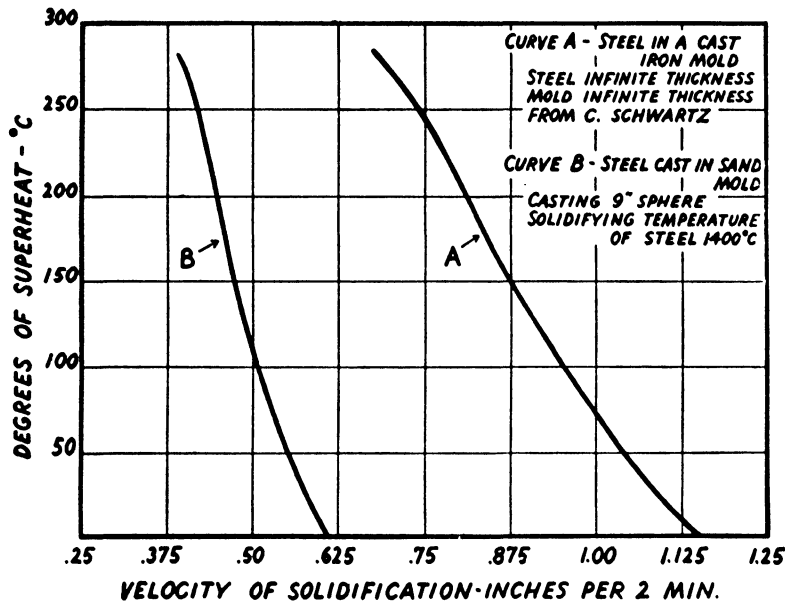
Fig. 68.—Effect of superheat on the velocity of solidification. (*Briggs and Gezelius.*[1])

Reentrant angles are formed at joining sections and at places of abrupt changes of thickness. In such angles the rate of skin formation lags because of insufficient cooling capacity of the mold. Figure 57 shows the effect of reentrant angles on the rate of skin formation. The effect of change in the size of joining sections on the rate of solidification is also illustrated. The volumes, surface areas, and ratio of surface area to volume of each cylinder are as follows:

Cylinder		Volume, cu. in.	Area, sq. in.	Area-volume ratio
Diameter, in.	Height, in.			
2	3	9.42	21.98	2.20
4	3	37.68	47.06	1.25
6	3.5	98.94	81.65	0.82

Castings that were bled at 5 sec., 1 min., and 2 min. show no difference in linear solidification in the three sections, even though the volume of each section is considerably different and the ratio of surface area to volume decreases rapidly.   After 2 min. this is no longer true, since the corner effect, owing to the fact that the heat is conducted away from one corner more rapidly than the other, has overshadowed the solidification perpendicular to the cylinder walls.

Thus the characteristics of the mold determine the length of time it takes before solidification is complete.   As the rate of cooling decreases, the time required for the steel to pass through the freezing range increases, and the velocity of solidification is altered.   The shape and size of the mold also govern, to a certain degree, the temperature gradient throughout the steel casting, for in small sections the chilling effect may bring about complete solidification, whereas in large molds solidification will proceed for considerable periods of time after the chilling action has been exhausted.

Effect of Superheat on Rate of Solidification.—The initial rate of skin formation is decreased by pouring castings with superheated steel.   The excess heat must first be extracted from the superheated steel before solidification can begin.   The effect of superheat on the solidification of ingots has been calculated[11] and studied experimentally.[3]   The effect of superheat on the solidification of steel in a sand mold is shown in Fig. 68.   For comparison a curve is reproduced to indicate the rate of solidification of a steel ingot.   The effect of superheat on the velocity of solidification of steel is not so great in a sand mold as it is in a cast-iron mold.

Effect of Temperature of the Mold on Rate of Solidification.—The rate of solidification of a metal in a mold depends, to a certain extent, upon the temperature of the mold prior to pouring.   It has been calculated[11] for steel ingots that heating the cast-iron mold is equivalent to superheating the steel.   No information on the rate of solidification is available as to quantitative values resulting from using heated sand molds.   It is evident, however, that the rate of solidification is less in molds heated prior to pouring.

Certain parts of a mold may become unduly heated through pouring the mold; hence the rate of skin formation does not proceed uniformly all over the castings, even though there is no corner effect or reentrant angles; but at gating and feeding centers it proceeds more slowly.

Effect of Fluidity on the Rate of Solidification.—The fluidity of the steel has no effect upon the rate of solidification of the steel, even though some foundrymen attempt to justify such a relationship.   Some producers of steel castings picture steel solidifying as large areas in a casting and assume that as it solidifies, it passes through a pasty or mushy state

of low viscosity.   In discussing this point Andrews[13] has stated that a determination of the freezing point of a steel will show that the greater mass of the steel becomes solid at a constant temperature, even though there may be wide differences between the liquidus and solidus.

Furthermore, Desch[7a] has stated that there is no evidence for the assumption that liquid steel becomes very viscous or passes into a pasty stage since the viscosities of molten metals are low, even close to the melting point.   The viscosity of iron and steel is only a small multiple of that of water, and the kinetic viscosity (viscosity divided by density), on the basis of which the effects of stresses due to viscosity in modifying the existing motion of a fluid may be compared, is considerably lower than that of water.   In castings that have been bled late in their solidification, there is porous steel adhering to the walls of the cavity.   This condition is caused by the slow transition from the liquid to the solid state and consists of an aggregate of crystal skeletons, continuous in the sense that the projections of adjacent skeletons touch one another or have grown together but contain liquid steel in the interstices.

CONTRACTION (SHRINKAGE) CAVITIES

It has previously been considered that in the solidification of steel castings the solidification proceeds from the surface inward, and as it proceeds, there are three types of contraction—liquid, solidification, and solid—which take place simultaneously in different parts of the casting. Thus, when the casting has completely solidified there will be a cavity at that place that was last to solidify, unless certain measures are taken to prevent its formation.

It was shown in Chap. V that carbon steel as normally used for castings solidifies in a range from 2625 to 2700°F., and that in going from the liquid condition to the solid state, for a 0.35 per cent carbon cast steel, there will be a 3.05 per cent volume contraction[14] as shown in Fig. 42. If the casting is not fed from some molten reservoir, a cavity will form in that portion of the casting that is last to solidify or in any portion that is cut off from the supply of molten metal.   The volume of such cavities depends upon (a) the mass of the solidifying metal, (b) the volume changes, and (c) the casting temperature.   Part (c) is really a part of (b), since the extent to which casting temperature influences contraction is governed essentially by liquid contraction (see Chap. V).

Contraction cavities, or shrinkage cavities as they are more commonly called, may appear in several shapes.   They extend from large, elongated shapes to micro-cavities surrounding grains.   The size of the cavity is governed somewhat by the section being solidified.   Figure 69 shows a shrinkage cavity uncovered by sawing a section in half.

Uniform sections that solidify often have a small amount of shrinkage along the axis of the section. This type of cavity is known as "center-line shrinkage" or "center-line weakness." An illustration of this type of cavity is shown in Fig. 70. Center-line weakness, or center-line shrinkage, is often encountered because of the lack of temperature gradients along uniform sections. Center-line shrinkage[20] is a porous condition of metal consisting of a labyrinth of cavities whose geometric character changes with viewing position. In vertically cast sections
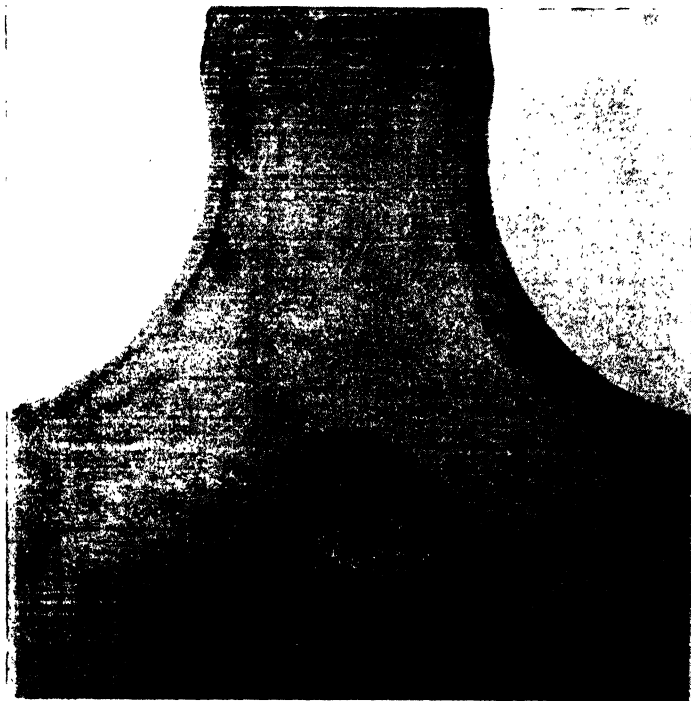
Fig. 69.—Shrinkage cavity in an unfed T section. Section sawed in half to uncover cavity

viewed edgewise it consists of a series of nested concave-sided macrovoids whose vertexes point away from the feed head and lie centrally located between the cast surfaces.

The cavities consist of tiny crevices with dimensions often on the order of a fraction of a millimeter, with irregular and ragged outlines and with nonoxidized walls. The defect is found in the last solidifying portion of a uniform section. It can be corrected in two ways:

1. The establishment of a definite temperature gradient along the uniform section in the direction of the feed riser.

2. The positioning of a number of feed risers along the section at short and regular intervals of spacing.

When a shrinkage cavity extends to the surface of a casting, it is known as a "pipe." Also, all shrinkage cavities in risers or feed heads are spoken of as pipes. In many instances the word "pipe" is used synchronically with shrinkage cavity.
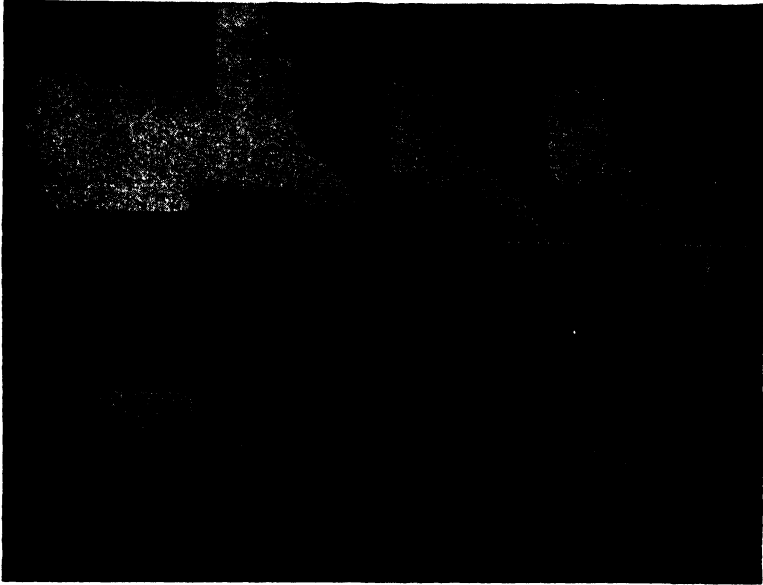
FIG. 70.—Center-line shrinkage. × ½. (*Courtesy of Naval Research Laboratory.*)

FIG. 71.—Micro-shrinkage in a steel casting. × 150. (*Portevin.*[20])

Interdendritic pipes (micro-shrinkage) are tiny cavities with dimensions often on the order of a fraction of a millimeter, with irregular and ragged outlines. They occur at the intersection of convergent dendritic

directions where dendrites growing in various directions meet. The contraction at the final point of solidification produces the microscopic cavities. Tensile-test specimens prepared from areas containing micro-shrinkage and center-line shrinkage will show a considerable decrease in elongation and reduction of area. An illustration of micro-shrinkage (interdendritic pipe) is shown in Fig. 71.

Hot-spot Formation and Steel Casting Design.—Shrinkage cavities occur in steel castings because of poor design, since they are usually found in sections that must be fed through smaller sections. If there is no way of placing an outside reservoir of metal on the section, a cavity will result, since the section is unable to draw liquid metal from other sources when those inlets have completely solidified.
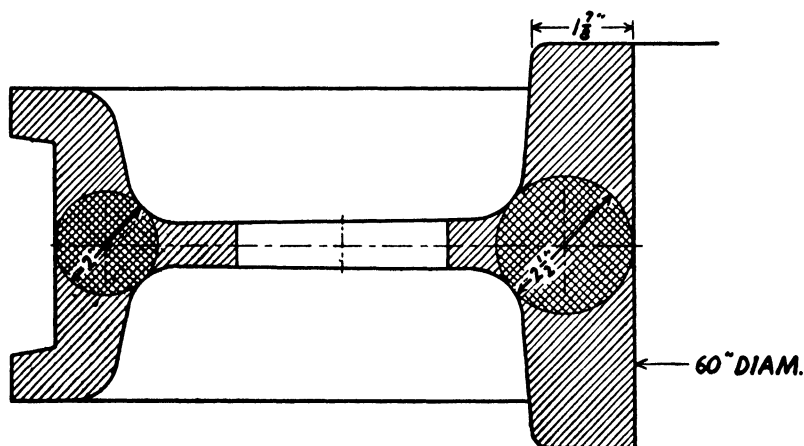
FIG. 72.—Cross section of a crane traveling wheel.

When in use, castings with shrinkage-cavity defects, under vibratory stress aided by stress centralization, may develop cracks extending from the cavity to the casting face. If the casting is operating under pressure, leaks may develop.

The problem of removing shrinkage cavities results in the elimination of the hot spots in castings. Isolated hot spots are a source of great trouble in the manufacture of steel castings. Hot spots are sections of extra mass, found for the most part at positions of joining sections.

A simple illustration of hot spots is given in Fig. 72, of a crane travers-ing wheel. By the method of inscribed circles the hot spots of the casting can be located. If these hot spots can be fed only through the overlying 1⅞- and 1-in. sections, it is certain that when the casting has solidified, there will be a cavity somewhere near the center of the inscribed circles, since the smaller sections solidify first. If, however, the hot spots are in a location where the junctions could be fed directly by an outside

reservoir rather than attempting to feed through the smaller sections, then the sections at the hot spot would be sound and contain no cavity.

There are numerous occasions, unfortunately, when joining sections such as shown by the crane cannot be fed by an outside reservoir of metal because of inaccessibility. There is no special problem in joining sections or hot spots that can be fed directly, and design features are not very important so far as safeguards against cavities are concerned; but sections that cannot be fed should be considered carefully by the designer and foundryman.

One of the most difficult features about designing to prevent contraction cavities is to know when and where feed heads may be placed, since their position is quite largely a matter of opinion among foundrymen. Castings may be molded contrary to any plans that the designer may have had. Thus the casting may better be designed with the supposition in mind that none of the metal conjunctions or hot spots can be fed from outside reservoirs, or the casting design should be reviewed by the foundry selected to make the casting.

There are only five ways in which sections may be joined. These may be exemplified by the following letters of the alphabet: L, T, V, X, and Y. All other modes of connection are merely modifications of the above possibilities. Designs in the form of these letters have been studied[15] in detail to ascertain the effect of the joining sections on cavity formation. Outside feeding reservoirs were not placed on the junction of sections.

L Section.—If the outside corner at the junction is maintained, an increasing radius at the inner corner will bring about an unnecessarily large defect. If the section is uniform throughout, a defect will be found if the inner radius is small. If the inner radius is increased but a uniform section is maintained, the shrinkage cavity will become smaller until it develops into center-line shrinkage. A solid casting is produced by designing the section at the junction slightly smaller than that of the arms and using an inside radius of $\frac{1}{2}$ to 1 in.

T Section.—The only way in which a T section can be designed to avoid a contraction cavity is to core a hole at the center of the junction of the two members. Depressions in the arm of the cross member reduce the defect markedly from that found in the common T design. Since the size of the cavity increases with the size of the radius, a radius of $\frac{1}{2}$ in. and not over 1 in. is recommended.

V Section.—A uniform section in the shape of the letter V will not be free from contraction cavities. As in the case of the L design, a slightly reduced section at the junction of the members is necessary for a homogeneous section. In designing unfed joining sections in L or V shapes, it is suggested that all sharp corners at the junction be replaced

by radii so that this section becomes slightly smaller than that of the arms. The inner radius should not be less than 1 in.

X Section.—If the X section is fed only through the arms, it cannot be designed free from contraction cavities. However, a cored hole at the center of the junction is helpful. Sections that join in an X should be designed so that the two arms are offset considerably. This will permit the foundryman to use chills effectively to produce a solid section.

Y Section.—The Y section is much like the X section and is difficult to design to provide freedom from shrinkage cavities. A wide radius should be used at the junction of the upraised arms. In designing any joining section, it is suggested that all sharp corners at the junctions be replaced by radii. In the case of unfed T and X sections, these radii should not be large. All large, isolated masses represent poor design. A heavy section should either be cored or carefully studied in order to ascertain whether it could not be made lighter.

It is desirable to emphasize two things: First, the sections that have been discussed were not separately fed; had a reservoir of liquid metal been attached to the sections, they would have been free from contraction cavities. For this reason, the foundryman will go to considerable trouble to feed isolated joining sections. Second, even though the designer is unable to design a section that, if not fed, will be free from contraction cavities, he can design it so that it will have the smallest cavity possible. With such a design the foundryman can apply controlled directional solidification, padding, external or internal chills, to produce a homogeneous section.

CONTROLLED DIRECTIONAL SOLIDIFICATION

Any liquid metal cooling in a cavity formed in a refractory or metal mold will solidify progressively from the mold-metal interface toward the center of the cavity (Fig. 73*A*), and, in addition, solidification will proceed along the cast member in the direction of the increasing temperature gradients produced in pouring the casting. These temperature gradients along the cast member appear in both the metal and the mold and are determined by the shape of the mold cavity and the method of gating and heading (Fig. 73*B*). They are, therefore, controllable, whereas the foundryman has no control over the direction of the temperature gradients producing progressive solidification. It is evident that all shrinkage cavities in castings are the result of adverse temperature gradients or uncontrolled directional solidification.

Only when the temperature gradients within the metal and the mold are such that the feed heads are the last portion of the casting to solidify and thus can furnish liquid metal to fill the voids formed during solidification—only then can a "sound" casting be produced. Since in most

castings, desirable temperature gradients can be produced by using tapered sections, proper methods of gating and heading, and properly placed chills, it follows that controlled directional solidification (Fig. 73*C*) is the governing factor in the elimination of cavities due to shrinkage.

Temperature Gradients.—In Fig. 59, showing temperature studies on a 9-in.-diameter sphere cast with steel, it may seem that there is a difference of 350°F. between the center of the sphere and the mold-metal interface during the first 20 min. of cooling. Castings accidentally bled as the result of a breakout near the gate have shown by their remnant shells an envelope of a thickness considerably greater at the upper levels
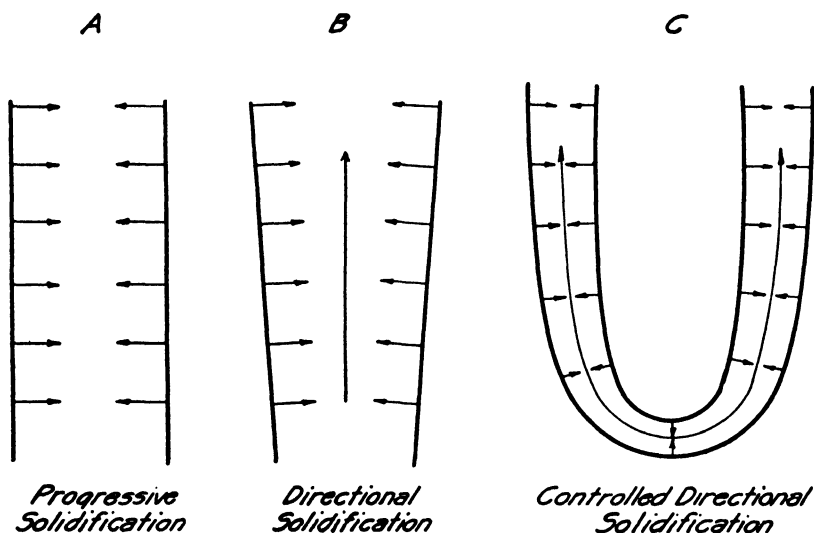
Fig. 73.—Solidification in castings.

of the casting than at the lower levels close to the ingate. This demonstrates the effectiveness of the mold as a coolant of the molten steel passing over it.

An idea of temperature-gradient distribution in a mold can be obtained by the design of Fig. 74. This cross section of a portion of a casting could be a part of a cylinder or a box section. The casting is filled with molten metal through a bottom gate. As the metal rises in the mold, it drops in temperature because the cold walls of the mold take up some of the heat. The metal that arrives at the top of the riser is relatively cold metal, since it has lost heat to the mold throughout its journey; the metal at the bottom of the mold near the gate is relatively hotter metal.

All the metal in the mold flowed through the gate and over that portion of the mold adjacent to the gate. Despite the low thermal con-

ductivity of sand, the mold takes heat from the metal in transit.   Therefore, the mold near the gate is relatively hotter than the mold adjacent to the riser.   Temperature gradients are therefore established in the metal and in the mold, with the highest temperature established in both metal and mold near the gate.

It has been observed by Batty[16] that temperature gradients of 125 to 150°F. can be established in the bottom pouring of small- to medium-sized castings, by normal fast pouring-practice methods when bottom gating is used.   If slow pouring is used, a temperature gradient of 150 to 200°F. may be found between the top and bottom of a casting.

The section shown in Fig. 74 will solidify at the point *PP* because the initial skin formation was not so great in the lower half of the section as in the upper portion.   In fact, the adverse temperature gradient established by the method of pouring has resulted in the formation of a hot spot in the casting section near the gate.   Since the section at *PP* has completed its solidification and the gate is solid, there is no way to feed the lower portion of the casting; consequently, shrinkage cavities develop when this portion completes its solidification.
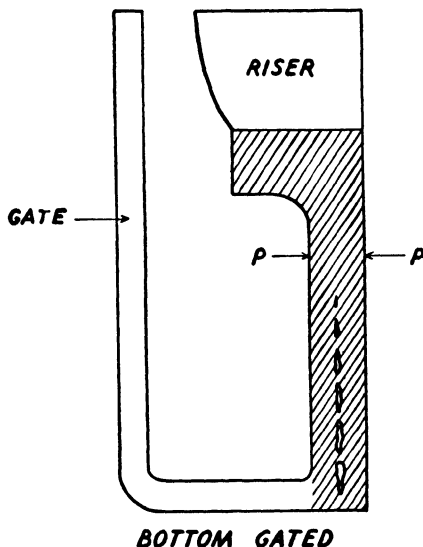
Fig. 74.—Cross section of a portion of a casting.

Temperature gradients in a mold can be controlled during the pouring of the mold by
1. The positioning and proper functioning of gates and risers
2. The speed of pouring
3. The use of hot metal
4. Increasing casting section by the use of padding sections
5. The use of molding materials of different thermal characteristics
6. The use of metal chills

Positioning of Gates and Risers.—The establishment of temperature gradients in metal and mold by the positioning of gates and risers is discussed in detail in the next chapter.   In general, this control is effected through the use of one of the following:
1. Partial- or total-reversal mold manipulation
2. Top pouring or top gating of castings
3. Step and multiple gating of castings

4. Top filling of risers

5. Feeding by the use of blind risers and open risers

The methods of partial- or total-reversal manipulation of the mold have been thoroughly described by Batty.[16,17]   Batty took advantage of the adverse temperature gradients found in the bottom-pouring method of gating; by rotating the mold in the vertical plane from 30 to 180 deg., he was able to place hot metal in the most advantageous gravity feeding positions.

Many castings have been produced by the partial-reversal method wherein the angle of rotation is from 20 to 45 deg.   This procedure is adaptable to large castings as well as small ones.   The method was used
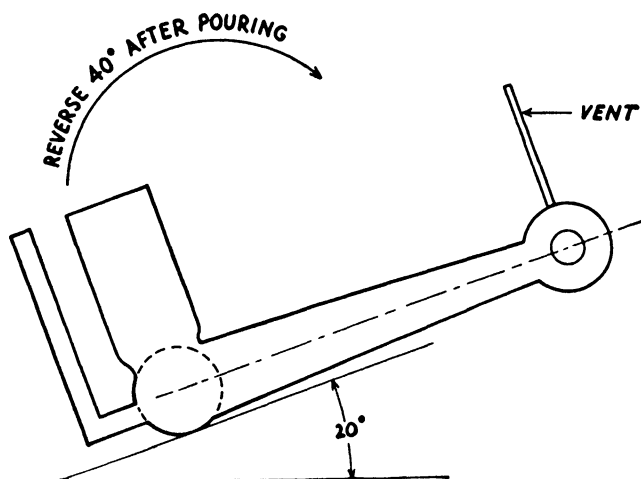
FIG. 75.—Cross section of a 20,000-lb. stackless anchor.   Partial-reversal mold manipulation.   (*Duma and Brinson.*[18])

by Duma and Brinson[18] to produce a 20,000-lb. stockless anchor, a cross-sectional sketch of which is shown in Fig. 75.   The procedure of mold manipulation is to establish a temperature gradient along the shank toward the riser.   The mold is tilted at an angle of about 20 deg., so that the ingate and riser are located at the low end of the mold, as shown in Fig. 75.   After pouring is completed and the vent sealed, the mold is reversed through an angle of 40 deg. to bring the riser in an elevated position in relation to the casting.

The gating system is devised in the partial-reversal method to introduce the metal quietly into the lower levels of the mold without producing undue turbulence.   The gate and riser areas of the mold are progressively heated as the mold is filled, while the metal is progressively cooled by transit through the mold.   It is not suggested that the first metal that enters the mold is ultimately driven into a position at the upper end of the mold near the vent, but it is affirmed that at the instant the mold is

filled, the average temperature of the metal across the section of the casting most remote from the ingate is appreciably less than that of the average temperature of the section of the metal immediately surrounding the ingate. It is conceivable that between these two extremes there is a fairly regular temperature gradient. It has been estimated that in the case of the anchor shank, a temperature gradient of approximately 700°F. was produced.
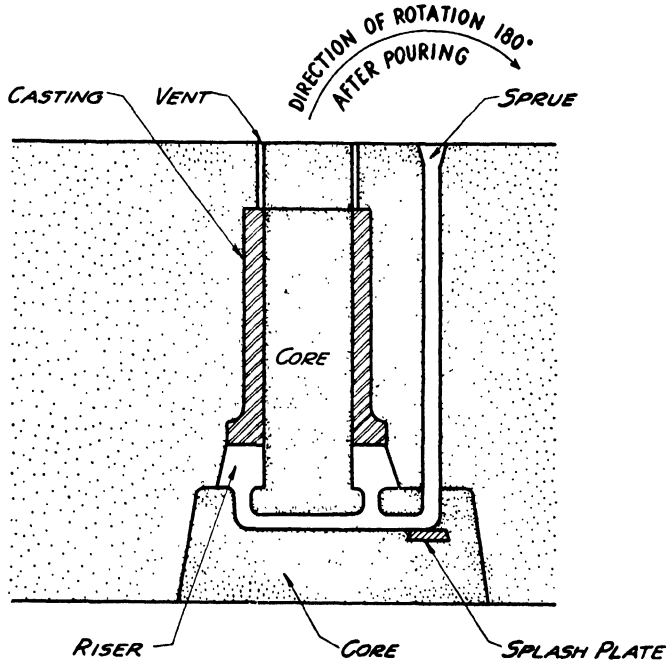
FIG. 76.—Cross section of a liner casting. Total reversal mold manipulation. (*Batty*.[17])

The mold-manipulation principle may be extended to the full use of bottom gating and top feeding by the use of the 180-deg. reversal plan as shown in Fig. 76. The most favorable temperature gradient in both metal and mold may be attained by a system of operation that has been called by Batty[16] the "total-reversal method." Within limitations prescribed by the available equipment, this method is productive of very definite economies on certain types of casting, and at the same time it promotes a degree of integral soundness of structure unattainable by ordinary methods.

It is very difficult to feed the uniform cylinder section of the lines of Fig. 76 with orthodox methods of gating and feeding. Under the total-reversal plan of pouring, the metal enters the risers first and then rises in the casting. A very pronounced temperature gradient, estimated

to be about 200°F., is established, which, after mold reversal, brings the riser into its proper position to deliver feed metal by gravity.

With the casting in the reversed position after pouring, the lower levels of the metal are relatively cool and are in contact with the cool mold, while the upper levels—particularly the riser—will be maintained in a fluid condition for a considerable length of time by reason of the heat of the mold with which they are in contact. The positioning of cold metal-cold mold with regard to the hot metal-hot mold permits the solidification of the uniform section without shrinkage cavities or center-line weakness.

A study of temperature gradient in a 3-in.-diameter section has been made by Taylor and Rominiski.[19]   They bottom-gated a 3-in.-diameter
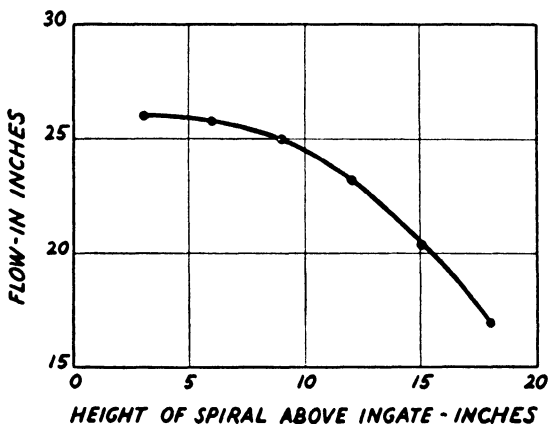
Fig. 77.—Temperature distribution in a bottom-poured 3-in.-diameter cylinder casting. (*Taylor and Rominski.*[19])

vertically molded casting with a 1-in.-diameter sprue and took fluidity spirals off from the cylinder casting at 3-in. intervals. Figure 77 shows the fluidity values obtained. When these fluidity points are correlated in terms of temperature for a carbon steel, a temperature gradient of 85°F. is indicated between the bottom and the top of the casting. This pronounced temperature gradient in the molten steel is indicative of what may be expected in uniform sections that are bottom-gated. It should be noted, however, that as a section increases in breadth, the temperature gradient will decrease.

Temperature gradients in metal and mold may be brought about by gating and feeding operations other than partial- or total-reversal mold manipulation. By top pouring or top gating of a casting it is possible to circumvent the adverse temperature gradient normally found in bottom gating of a casting. This adverse temperature gradient may be partially corrected by the top filling of risers. By the use of step or multiple gating of a casting, it is possible to reduce the hot spot in the

casting section near the gate since all the metal would not enter the casting through a single gate (see Chap. VII).

Blind riser feeding takes advantage of the temperature gradient established in sections by bottom pouring, by gating directly into a riser at or near the bottom of the casting. Since all the metal enters the riser prior to entering the casting, a thermal gradient is established that permits the riser to solidify last with feeding proceeding in the direction of the thermal gradient.
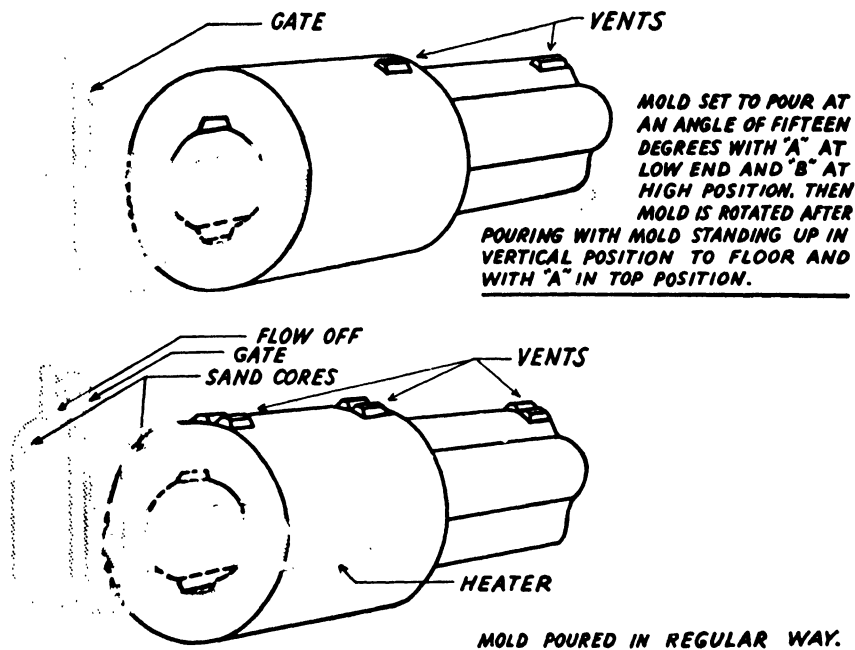
FIG. 78.—Steel casting previously poured by the partial-reversal method altered to the blind-riser-feeding practice. (*Courtesy of E. C. Troy, Dodge Steel Company.*)

The control of temperature gradients seems to be the most effective regulator of center-line shrinkage in castings. By the use of bottom pouring through blind risers acting under atmospheric pressure, it is possible to produce homogeneity in uniform sections that would normally show center-line shrinkage. This does not mean that the blind riser is more effective than the open riser in all cases; it merely means that a correct temperature gradient is necessary in both metal and mold in order to obtain the proper degree of solidity. Open risers may be constructed to feed equally well if the proper temperature gradients are established. In many cases it is more difficult to establish temperature gradients for the effective use of open risers. This condition has led to the common belief that blind risers are in themselves more effective. They are not

more effective, but they are more advantageous since they can be used with the normal bottom-gating methods to establish easily the required temperature gradients. Blind risers can be so situated and employed that they may not feed sections effectively.[20]

In certain cases the advantages of partial-reversal manipulation can be obtained by the use of blind risers with the atmospheric-pressure principle of feeding without going to the trouble of rotating the mold. Figure 78 illustrates a steel casting that previously was produced by the partial-reversal method and now is being poured on the horizontal through blind risers. Temperature gradients established by the blind-riser feeding were sufficient to the production of a sound casting without resorting to the more elaborate mold manipulation. Information as to the temperature gradient established in a particular casting by the two pouring methods has not been ascertained.

Speed of Pouring.—The loss of heat of molten metal as it traverses the mold is proportionate to the duration of contact of the molten metal with each mold increment and the ratio of the contacted mold area to the thickness of the metal stream passing over it. The heat transference from the metal to the mold will therefore be related to the speed at which the mold is filled. The rate of rise of metal in the mold cavity is instrumental in determining the magnitude of the thermal gradients in the casting. In order to overcome the adverse temperature gradients of bottom gating and top feeding, it has been the general policy of casting producers to fill the mold cavity as quickly as possible with metal as cool as possible. It must be conceded, however, that in many cases the application of this principle, even when carried to its ultimate practical limit, has failed to produce castings free from shrinkage defects.

The control of the rise of metal in the mold cavity has not been given the consideration it deserves. The ingate diameter and the speed of metal flow should be more carefully studied, for in this way it is possible to standardize conditions so that results will be uniform.

It is the policy under the present-day conditions to inspect by non-destructive testing methods pilot castings, in order to observe—among other things—their freedom from shrinkage cavities. It has been stated that even though a pilot casting is produced free from such defects, it is not an absolute certainty that other castings produced in a similar manner will not contain cavities inasmuch as the process of manufacture is not sufficiently controlled. Metal pouring temperatures vary too widely, and variable temperature gradients are established by inability to standardize on the rate of rise of metal in the mold. The slower the rise of metal in the mold, the greater will be the temperature gradient produced. The change in temperature gradients caused by these two items is sufficient to result in variations in cavity formation. It is predicted that a great deal

more study will be given to the rate of rise of metal in a mold in the near future, because it is a variable that directly affects thermal gradients in the metal and mold and is an important consideration in controlled directional solidification.

Use of Hot Metal.—The use of hot metal together with slow pouring is a combination that gives rise to pronounced temperature gradients within the metal and the mold. Admittedly, high-temperature metal and slow pouring would be faulty in most instances—even dangerous—under conditions of orthodox gating and heading procedure. If, however, thought is given to directional solidification, then the use of hot metal in attaining this end is advisable. The limit to the practicable use of hot metal is related to the effect of the metal upon the mold and the gates. High-temperature metal or metal rushing into a mold at high velocity tends to promote defects in steel castings if the molding materials are not thoroughly competent to withstand the thermal and physical attacks of the metal.

The use of hot metal can be worked into a controlled system of pouring by the examination of pilot castings poured with steel of different temperatures. With the rise of metal in the mold set at a definite figure, a pouring temperature range of 100°F. could be investigated with pilot casting at the top and bottom of the range. All castings so investigated could be given preference in the pouring position of each heat that would be equivalent to the temperature range tested by the pilot castings.

Castings with high-temperature metal will increase the amount of liquid contraction that must be provided for, by feed metal. High-temperature metal also reduces the initial skin formation. It has been stated that high-temperature metal is responsible for producing a coarse structure in the casting and for permitting the occurrence of serious segregation. Such claims have not been proved.

Padding for Directional Solidification.—It has been emphasized that a uniform section will complete its solidification throughout its length at the same time, with the formation of central shrinkage cavities called "center-line shrinkage." If it is possible by gating, risering, and pouring methods to develop pronounced temperature gradients in the uniform section, the section can be made to solidify directionally toward the feed head.

In some cases no pronounced temperature gradients are established, or adverse gradients are established in the uniform section. To prevent the latter and to allow for a positive feeding direction toward the riser, it has been found advisable to pad the uniform-thickness section. The padding will take the form of a wedge addition to the section so that there will be no increase in the section at the point farthest from the riser, but above this point additional width will be given to the section in the form

of a taper or wedge that increases in thickness toward the riser. Padding is therefore a tapered disposition of metal on the walls of castings, the taper increasing in the direction of the feed heads. When correctly applied to casting members, it is productive of uniformly dense and internally sound cross sections.

Steel castings can be designed with the taper section as an integral part of the section. They are usually not designed as such, because the designer does not know how the foundryman will produce his casting, and any tapering of sections that the designer may provide might produce temperature gradients opposing those that the foundryman finds essential in securing a good casting. If, however, the designer maintains fairly uniform cross sections throughout the casting, the foundryman can apply
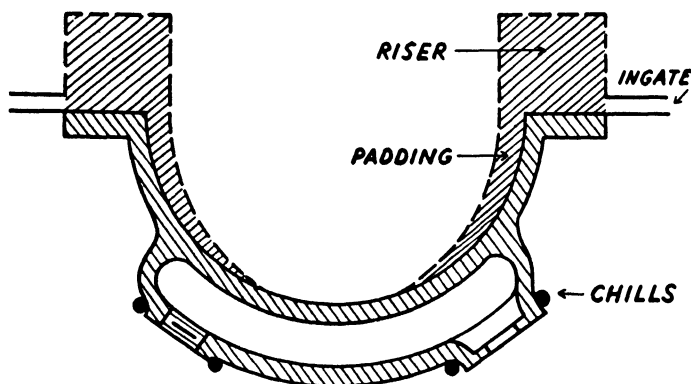
FIG. 79.—Application of padding to turbine casing. (*Briggs, Gezelius, and Donaldson.*)[15]

the padding to produce the temperature gradients that he desires. The trouble with this is that the casting buyer may require that the padding be removed from the casting. If, however, the foundry engineer and the designer collaborate on the design of the casting, such items as padding can be incorporated in the finished design prior to the pattern preparation.

Padding may be removed from sections by machining or by flame gouging and washing, but the operation is time consuming and adds to the cost of the casting, though this must be done when saving weight is an important consideration. In large castings this operation is much less costly than to scrap the casting and replace it with another because of the presence of shrinkage cavities.

Padding may be left on the casting, with approval of the purchaser; and since castings are generally sold according to weight, the extra weight due to the padding can be deducted in determining the sales weight. It is best, of course, to have collaboration with the designer in order to produce the proper casting design based on the fundamental principles of

the solidification and contraction of steel, which govern the shape of castable structures.

An illustration of how temperature gradients produced by tapered sections can ensure correct solidification is given in the sketch shown in Fig. 79.   If this casting is poured flange up, the sections tapered as shown, and a collapsible core used to forestall hot tears, it is comparatively easy to produce a sound casting.   It should be noted that the padding on the cylinder bore is brought down below the intersection with the steam chamber so that the steam chamber can be fed directly.   The casting
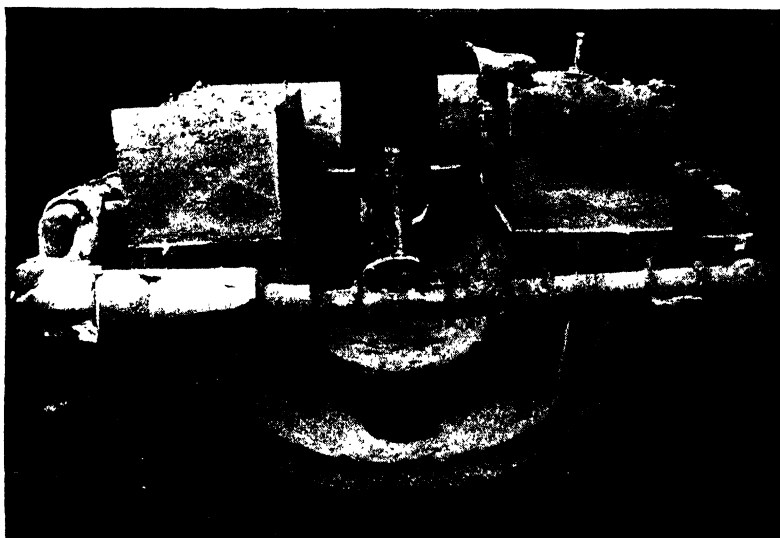
FIG. 80.--Position of gates and risers on turbine casing.   (*Duma and Brinson.*[18])

was gated at the base of the riser, as shown in Fig. 80, so that the last metal to enter the mold would enter the riser.   By placing external chills around the parts of the steam chamber as shown in Fig. 81, it can be readily seen that solidification proceeds from the lower portion of the casting toward the risers.   This direction of solidification has been controlled by producing favorable temperature gradients through the use of the method of gating, tapered sections, and chills.

Only one quantitative study[20] has been made as to the amounts of padding necessary in uniform sections to permit the section to solidify without center-line shrinkage.   In these studies, various thicknesses of rectangular plates of uniform and wedge-shaped sections were cast with feed heads at one end and a gate directly under the feed head.   These plates were radiographed and machined into various test specimens.

It is reported by Brinson and Duma[20] that center-line shrinkage is most noticeable in $\frac{1}{4}$-, $\frac{1}{2}$-, 1-, and 2-in. sections.   It faintly appears in

the upper half of 3-in. sections and is scarcely visible in the topmost metal of 4-in. sections. In the 4-in. sections the combined action of the strong end and side effects were sufficient to set up temperature gradients that permitted the section to freeze without the center-line shrinkage cavities. This gradient effect fades out in ratios of length to width to thickness of 8 to 8 to 1. The influence of increased height in sections susceptible to center-line shrinkage on the suppression of this shrinkage is negligible.
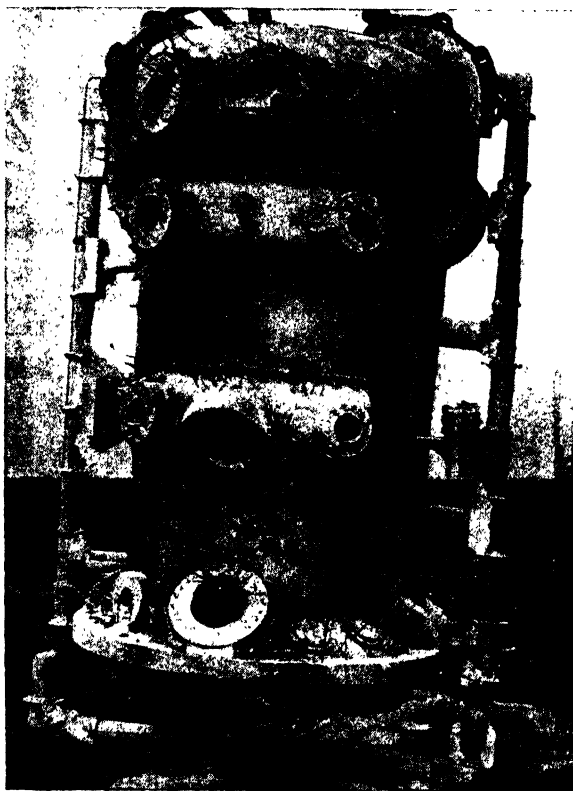
Fig. 81.—Position of gates and chills on turbine casing. (*Duma and Brinson.*[19])

There is apparently a marked tendency toward suppression of center-line shrinkage only in sections in which the end effect (bottom chilling and top heading) is strong.

Tests were made on uniformly thick sections molded at various inclinations in the mold. It was noted that the position of the section in the mold alters the geometry of the cavities but does not eliminate the shrinkage. The center-line shrinkage is brought closer to the cope surface when the section is horizontally cast. It was further pointed out by Brinson and Duma that the use of sufficient feed metal in the form of

large, truncated open risers (80 per cent of the weight of the casting) is inadequate not only of itself but even with the accompanying mold pre-heat to prevent center-line shrinkage.

The application of taper in increments of 1 in. per ft. to castings 1 in. thick and 12 in. high progressively decreases the amount of center-line shrinkage to the point of complete elimination when taper-padded with $2\frac{1}{2}$ to 3 in. of metal per ft. The degree of internal soundness of padded sections cast flat or at 45 deg. equals that of sections cast vertically.

It is very apparent, from the work by Brinson and Duma,[20] that the end and side effects of a section, as well as gravity, are effective in modify-
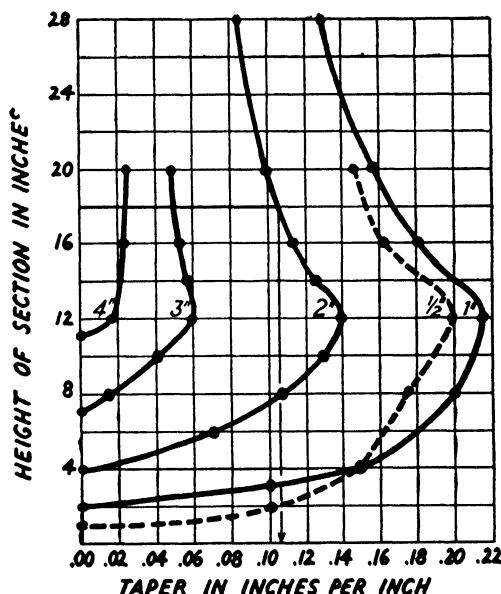
Fig. 82.— Padding required in inches per linear inch of section to obtain solid metal in sections of various height and thickness. This illustration shows that 1.91 in. (0.106 × 18) padding is required for a 2-in.-thick section 18 in. high. (*Brinson and Duma.*[20])

ing the padding requirements on short and long sections. The end and side effect makes its influence felt in long sections, and gravity is effective in short sections. The amount of padding in inches per inch increases for sections up to 12 in. in height, with the increase less for the heavy sections. Sections longer than 12 in. require less padding in inches per inch than do those of approximately 12 in., the decrease being greater for the thinner sections. Approximately one-half as much padding in inches per inch is required for 24-in.-high sections as for the 12-in.-high sections. This condition is shown in Fig. 82. It is obvious, from Fig. 82, that the volume of a section per unit of surface area per linear inch of either height or length is related to the degree of solidity of the metal. Therefore, the

gradient required to make solid sections is not a fixed constant but a fluctuating value varying with section height and thickness.

Center-line shrinkage in severe form has an adverse effect on test specimens taken from the section. When the test specimens are machined from the width of the section, the center-line shrinkage affects principally the elongation and reduction in area values. If specimens are obtained from the long axis of the section, the center-line weakness is more pronounced and the tensile strength and ductility are reduced by as much as 20 to 50 per cent. Properly padded sections show excellent properties in all directions. Comparatively lower notch sensitivity impact values
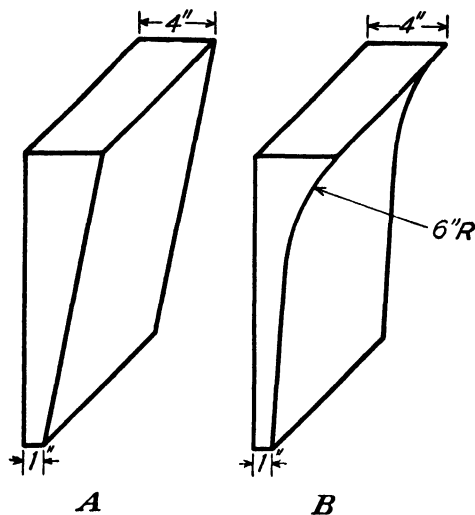
Fig. 83.—Types of padding. (*Brinson and Duma.*[20])

are found in specimens taken from sections containing center-line shrinkage than in solid specimens.

The geometry of pad design is not limited to straight-line tapers. Brinson and Duma show that besides the wedge type, there are other effective kinds of taper, such as curvilinear, broken rectilinear, or combinations of both. One such type is shown in Fig. 83*B*. This design is equivalent in effect to the 3-in. taper per ft.; yet it weighs only 27 to 30 per cent as much as the corresponding wedge design of taper. In padding sections it is suggested that the padding be applied to one side only. This side should be the easiest side to reach and preferably the side to be machined.

Even though the studies of Brinson and Duma are available, it is believed that many foundries that use padded sections do so empirically. The amount of padding generally applied is more often determined by trial and error, chiefly because the studies made by Brinson and Duma

were confined to small single sections, whereas castings consist of many sections. These many sections effectively change the distribution of the temperature gradients so that the excessive amount of padding, as demonstrated by Brinson and Duma, is not required.

If the foundryman uses the padding requirements recommended by Brinson and Duma for 1-in. sections, the amount of metal needed to pour the section is increased 250 per cent. The elimination of center-line shrinkage is very expensive if the padding must be removed before the casting is shipped. If, however, such padding can be incorporated as a design feature of the casting, the increased cost due to additional metal is not excessive for the type of casting where absolute solidity is required.

The use of padding for the elimination of center-line shrinkage on castings intended for high-temperature–high-pressure service and parts affecting the safety or military effectiveness of war machinery is undoubtedly necessary. However, in the case of structural castings that constitute the bulk of the steel castings produced, it is questionable if the properties of the castings as a whole are impaired by the center-line type of defect.

Center-line shrinkage probably exists to some extent in all parallel-wall castings. Steel castings for structural service are usually loaded as a beam. One surface will be in compression and the other surface in tension while the center line, which is the neutral axis, will not be stressed. A casting under these conditions is virtually unweakened by the presence of center-line shrinkage. In the discussion of Brinson and Duma's paper, C. E. Sims said that the railroad side frame casting, which is cast in about $\frac{5}{8}$-in. sections throughout, is operated under severe service conditions and is designed to stand a static load of about 500,000 lb. These castings have a splendid service record, even though center-line shrinkage is evident in the sections. Thus center-line shrinkage must be considered in terms of type of casting; and a casting should not be unqualifiedly condemned because it contains center-line shrinkage.

It has been established that specimens taken out of sections containing center-line shrinkage will exhibit a falling off in the properties normally expected for sound metal. If, however, the casting as a whole is considered, the resistance to stress concentration is quite different from that exhibited by specimens taken from the center of a section. In the axial loading of a casting the maximum stress is on the surface. Concentrations of stress are seldom located in a uniform section but are present at junctions of sections where center-line weakness is not prominent. Furthermore, the yield point of unpadded sections is only slightly lower than that of padded sections. The actual increase in the design strength due to padding is actually less than 5 per cent.

Extensive use of padding is not always necessary. Localized, highly

stressed areas can be well fed by padding or positioning of closely spaced risers. Castings, such as those shown in Figs. 79 to 81, that operate
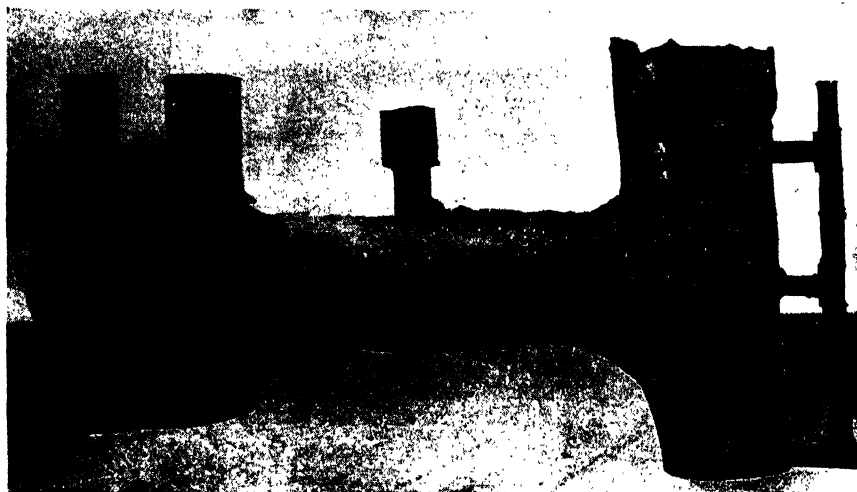
FIG. 84.—Two-arm strut as cast, showing gating, riser, and outside padding. (*Duma and Brinson.*[19])

under steam pressure and are required to be radiographically sound should be padded to prevent center-line weakness. One of the most serious factors contributing to leaky castings after machining is center-line shrinkage.
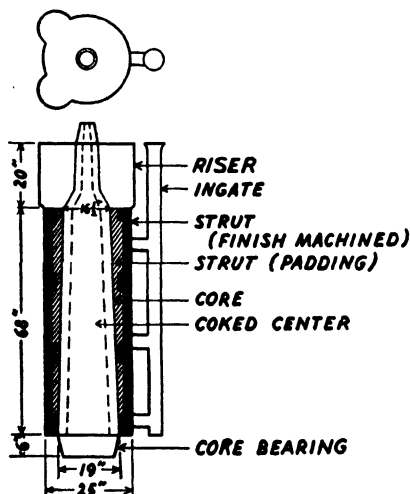
FIG. 85.—Design and internal padding of the bore of the strut. (*Duma and Brinson.*[19])

The two-arm strut, as shown in Fig. 84, is another casting that requires a high degree of solidity and is tested by radiographic methods. The bore portion is designed with uniform walls, except for additional isolated ring sections such as shown in Fig. 85. The step gate was used to correct the influence of an improper temperature gradient resulting from bottom pouring. A directional control of solidification toward the riser was assured by the internal padding of the bore from the bottom ring upward. A padding taper of 0.44 in. per ft. was used. Two external metal pads were also used, as shown in Fig. 84, situated directly over the junction of the bore and the arms.

By gating and padding in this manner and top-pouring of all risers, there was produced a pronouncedly favorable temperature gradient in both mold and metal that permitted the casting to solidify in a controlled directional manner. In the case with smaller single-armed struts the bore is cast horizontally. Padding is then applied as is shown in Fig. 86, to obtain controlled directional solidification.

Many foundries are reluctant to adopt the principle of padding. Other foundries have found that by the use of padding on large castings, the cost of the extra steel and its removal is appreciably smaller than the expenditure for casting repairs resulting from the use of an unpadded design. Furthermore, the defective portions of an unpadded casting
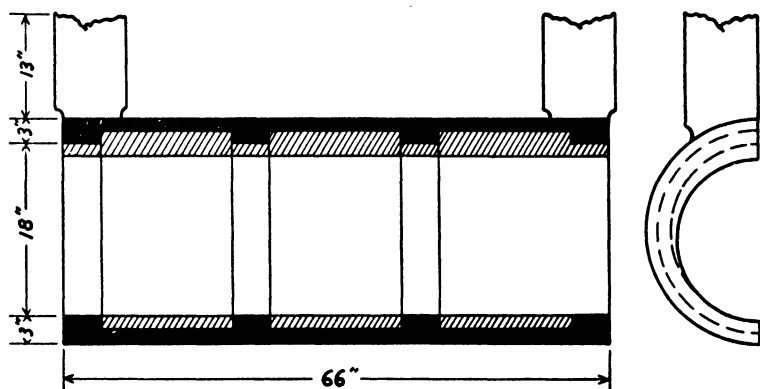
Fig. 86.—Padding in bore of horizontally cast strut. (*Duma and Brinson.* [18])

may be of such proportion that the entire casting must be replaced. The consequent loss of time and delay in production are also costly.

Varying Thermal Characteristics of Molding Materials.—It has been shown in Rate of Skin Formation, page 193, that molding materials have different heat conductivities. It is therefore possible to have controlled directional solidification in a mold by varying the cooling capacity of the mold in its different portions.

In selecting materials with a coefficient of heat diffusion higher than that of molding sands, it is advisable that the materials have the following requirements:

1. Sufficiently refractory so as to hold molten steel
2. Moldable
3. Should not produce casting defects, when suddenly heated by contact with molten metal
4. Permit clean, nice-appearing surfaces

The thermal properties of refractory materials alter with temperature. Since the temperature of the mold varies with the distance that the molding material is from the mold-metal interface and the elapsed time

after pouring, it is suggested that the practical way to determine the effect of various materials on solidification is to carry on tests under actual conditions.

The thermal properties of some refractory materials are given in Table XLI. From this table it appears that magnesite brick would be excellent material in controlling directional solidification, because its cooling capacity is from 2 to 2.5 times greater than that of an average molding sand. Magnesite used at reentrant angles exactly fulfills the cooling conditions required to attain the same solidification period in junctions as in the adjoining sections. The possibility of using magnesite

CROSS SECTION OF
UNIFORM SECTION CASTING IN FLASK
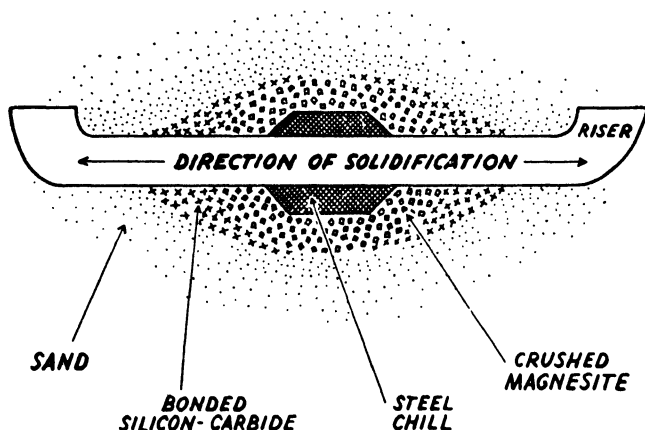
FIG. 87.—The use of molding materials of various thermal properties, to establish controlled directional solidification.

brick and bonded crushed magnesite has been confirmed by the writer with distinct success. The cooling ratio found experimentally with the magnesite used is about twice the cooling capacity of a normal grade of molding sand. It is reported by Chvorinov[10] that castings weighing over 20 tons were cast with large areas built in magnesite brick, without the least deleterious result and with the desired effect on the control of the direction of solidification.

Silicon carbide plastic materials used for lining and patching crucible furnaces can be used, giving a value intermediate between magnesite and molding sand. A further possibility is the use of a mixture of steel shot or turnings and a bonding material. Mixtures of over 50 per cent steel must be used in order to be effective, since the mixture of equal parts of steel and sand has only a slightly higher heat-diffusivity coefficient than that of sand.

A uniform section can be made completely solid by varying the molding materials in order to establish proper temperature gradients in the metal. The construction of the mold, in one instance, is shown in Fig. 87. The principle shown here can be extended in numerous ways, depending upon the casting design. Attention is directed to the fact that use cannot be made of a portion of the mold in green sand and another portion in dry-core sand; experimentation has shown[1,12,23] that the rate of skin formation under such conditions is practically identical.

Changes in heat flow of a mold can be obtained by varying the fineness of the molding material, such as making use of bonded silica flour, bonded alundum, and open-grain molding sand. The coarse, more permeable sand is placed at the position of desired heat losses, while the
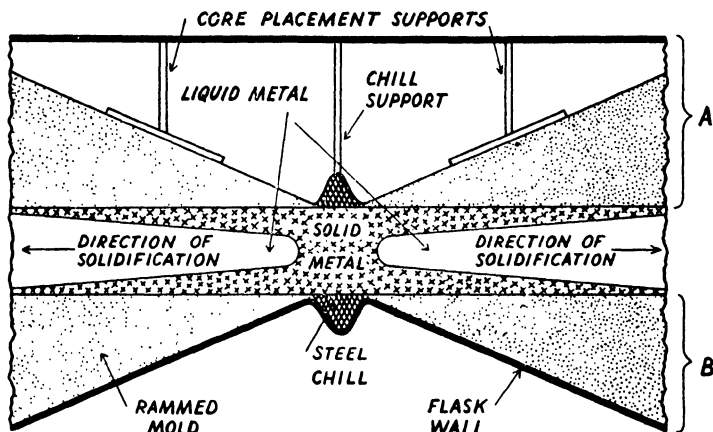
Fig. 88.—The use of various thicknesses of molding material to produce controlled directional solidification in a uniform section.

fine, low-permeability material is used at positions of desired low heat conductivity.

In nearly all cases in the production of steel castings the molding material extends some distance (from several inches to several feet) back of the mold-metal interface, depending upon the choice of flask equipment. Thus, from the standpoint of heat dissipation through the molding sand during the solidification of the casting, it may be considered that the mold is usually of infinite thickness, for several inches away from the mold-metal interface, the sand will not become perceptibly heated during the interval of casting solidification. It has been shown previously that sand is a very good insulating material.

It is suggested that solidification could be controlled and directed toward feed risers if only a very small quantity of molding material were used in certain parts of the mold, while in other parts of the mold greater thicknesses of molding materials could be used such as illustrated in Fig.

88. In view A the mold consists of assembled cores, while in view B the flask is constructed particularly for the individual job. The use of a metal chill is advisable in establishing the point of initial solidification.

The Use of Metal Chills to Obtain Controlled Directional Solidification.—Solid metals will abstract heat from liquid steel at a rate greater than molding materials. The most common metals used in the steel foundry for this purpose are rolled or cast steel and cast iron. When these metals are used for the purpose of extracting heat, they are called "chills." The effect of chills when placed in the mold at the mold-metal interface is to act to increase the initial skin formation. In this way
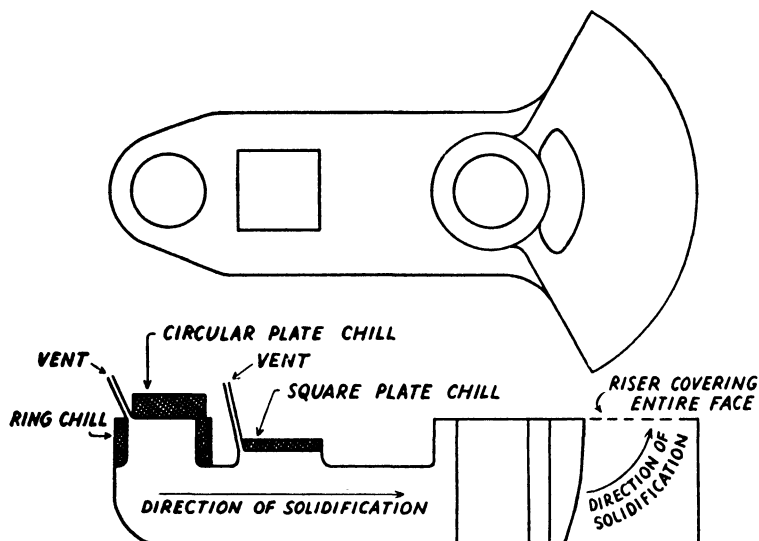
Fig. 89.—The use of chills to promote controlled directional solidification in a 250-lb. steel casting.

chills are able to make ineffective those hot spots at section junctions or at positions of isolated masses that normally produce defects.

Pads and bosses can be rendered ineffective in forming hot spots by the proper placement of chills. An example of this is shown in Fig. 89, in connection with the production of a 250-lb. casting. The ring chill, produced in one or two pieces, and the plate chills promote a heavy initial skin formation in the boss, which permits this portion of the cast cross section to solidify first. Solidification proceeds in the direction of the arrows, which is the direction toward the feed riser. Since the gate was constructed to enter at the base of the riser, a temperature gradient in metal and mold was established in the direction of the riser.

Figures 81, 87, and 88 show the use of chills to establish positions of solidification. It is usually beneficial to direct the solidification away from the chill area, as shown in Figs. 90 and 91. The casting illus-
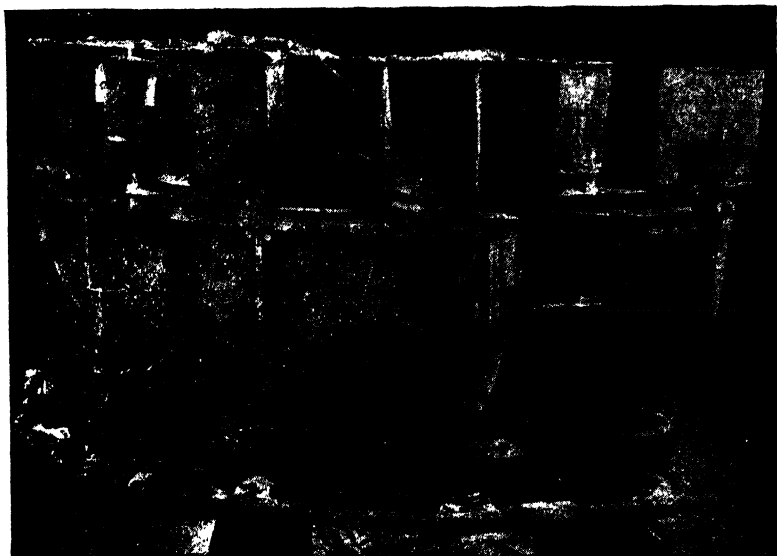
FIG. 90.—The use of chills in the production of a low-pressure turbine casing.
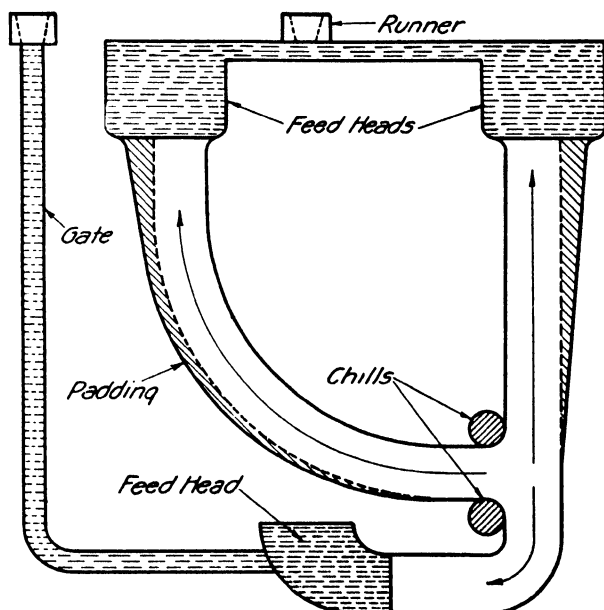
FIG. 91.—Cross-section view through the turbine casing of Fig. 90.

trated in Fig. 90 is a low-pressure turbine casing.   Attention is directed
to the chilling used at intersections and to the system of gates and risers.
Figure 91 is a sketch of the solidification details.   The junction of the
Y section is impossible to feed from feed heads located on the uniform
sections; therefore, this position was selected as the position where solidi-
fication would be completed first.   This was accomplished by the use
of chills at the junction of the section.   Padding was placed on the uni-
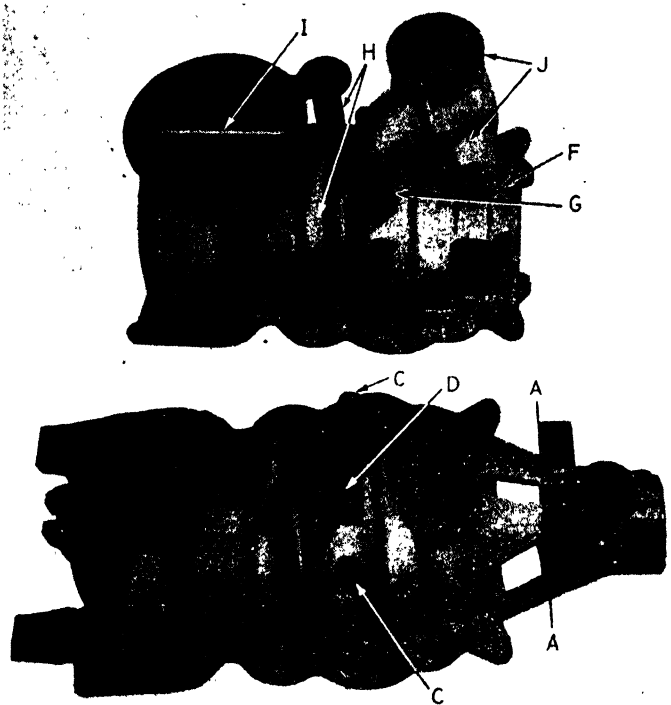
Fɪɢ. 92.—Model of intermediate-pressure-turbine cover and base.   (*Briggs, Gezelius, and Donaldson.*[15])

form sections so that they could solidify directionally toward the feed
heads.   The lower member feeds toward the blind riser because of the
establishment of gates into the risers, thus producing a very active tem-
perature gradient in the mold.   The entire casting was thus able to
solidify directionally and in a controlled manner.

In some castings it may be advisable to use indirect chills.   The term
"indirect chills" applies to metal chills that are not exposed directly to
contact with the molten metal but are protected by a layer of molding
sand or molding material.   The effect of an indirect chill is therefore
retarded and diminished, according to the thickness of the protective

layer and its heat conductivity. Employment of such indirect chilling methods may be advisable in establishing the direction of solidification.

The subject of chilling is treated from a more quantitative point of view under Chills and Chaplets (page 228).

Models as an Aid in Controlling Solidification.—One of the most efficient methods for studying the solidification of a casting and planning its directional control is to construct a model of the casting. Any con-
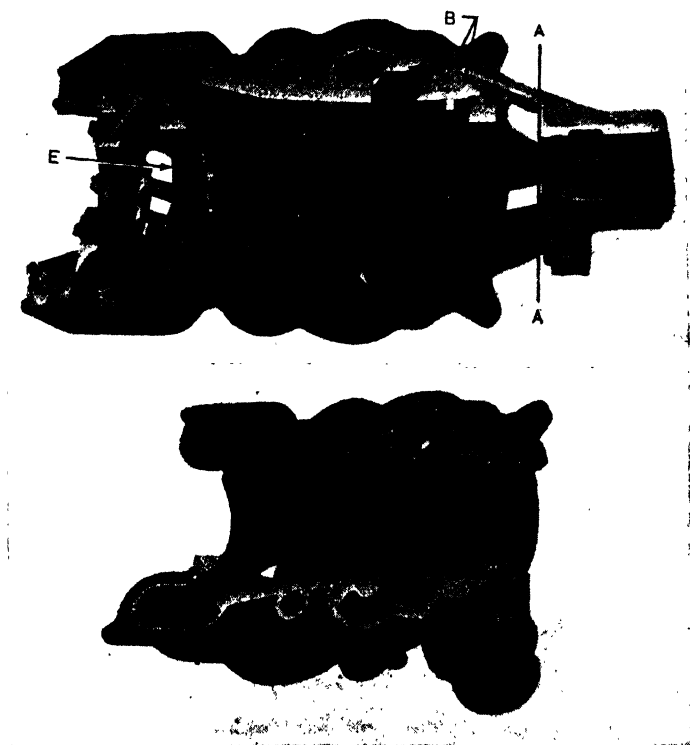
FIG. 93.—Illustration of section thickness of original design prior to addition of padding. (*Briggs, Gezelius, and Donaldson.*[15])

venient scale, such as 1 in. to 1 ft., is satisfactory. The model can be sectioned in several places, to show the section thicknesses. With a model of a complicated casting it is easy to visualize the method for joining sections, to observe the details of isolated heavy sections, and to plan for the position of padding and the placing of chills. The method of gating and feeding can be planned, and additions made to the model to show the gates and risers. The padding required can be produced to size scale, painted a color different from the casting, and attached to the model.

It is claimed by Brinson[21] that the construction of a model pays for itself, both as to time and cost, in the preparation of the full-scale pattern and production of the casting.

The model method of attacking solidification and contraction problems has been used to advantage by many foundries. Photographs of a

(a)

(b)
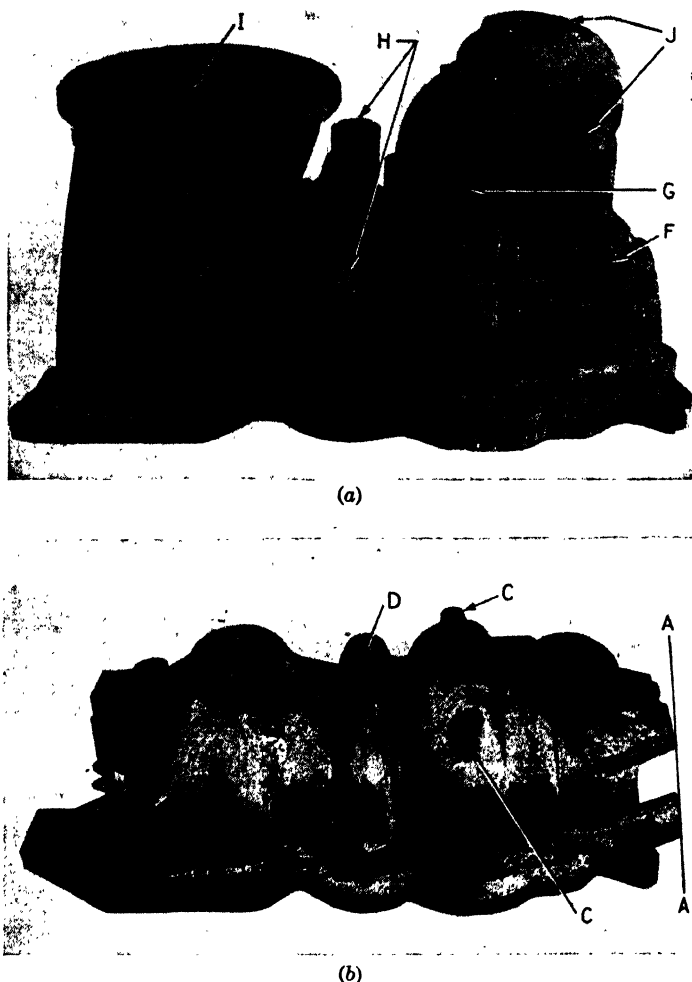
Fig. 94.—(a) Turbine cover, (b) base castings. (*Briggs, Gezelius, and Donaldson.*[15])

turbine casing,[15,21] produced after the design had been modified following the preparation and study of models, are shown in Figs. 92, 93, and 94. If the photographs of the models and the castings are compared, it will be noted that while the changes are minor in character so far as the

engineering usefulness of the casting is concerned, every effort had been made to obtain directional solidification and to simplify the production of the casting. In addition to the changes listed below, padding was provided to both castings, as shown for the base casting in Figs. 79 and 80. Additional changes made were as follows:

1. The base casting was separated at A and the high-pressure end was cast separately. This facilitated the production greatly.

2. The thickness of the gland-supporting web was increased and the cored "pockets" eliminated by changing the shape of the web.

3. The bolt bosses were tapered from top to bottom, to facilitate the feeding.

4. The thickness of the steam belt was increased and an opening provided for core anchorage. This opening was later closed by a plate welded on.

5. The radius at the junction of the flange and web was increased to eliminate a sharp change in section and permit feeding of the sections at the bottom.

6. The thickness of the web was increased and a larger radius provided to ensure proper feeding.

7. One outlet was cast separately and welded on.

8. The wall thickness of the steam belt was increased and the outlet was cast separately and welded on. Since the castings were to be used as a "right" and a "left," two openings were made to secure proper core anchorage. One of these openings was sealed by a welded cap; the other was welded to the outlet case separately. This permitted the use of identical cores for both the right and left castings.

9. The web was cast separately and welded on to the main casting. This web, if cast integrally, would very probably have caused hot tears, which would have had to be repaired by welding after considerable chipping.

10. The walls of the by-pass valve were brought straight down to the casing to eliminate sharp changes in direction of section and permit better feeding. The outlet was cast separately and welded on.

In the case of the turbine castings, the preparation and use of a model not only resulted in changes in design that facilitated production but also assisted in the planning and control of the direction of solidification. Models of castings aid the buyer and the design engineer actually to visualize the details of their casting. This is especially helpful for prospective castings of complex design. In many cases, after the purchaser has seen the model, he will be most agreeable to making design changes suggested by the foundry, for he can actually see how such changes will alter the appearance or the function of the casting.

CHILLS AND CHAPLETS

· Chills, or metal inserts, have been used in the production of steel castings for many years. These metal inserts are of two major types as to their function: (1) external chills, and (2) internal chills. External chills are placed in the mold walls at the mold-metal interface, while internal chills are placed in the mold cavity.

The chills used for steel castings are for the most part constructed of rolled, forged, or cast steel. In fact, all internal chills are manufactured from steel since these inserts become a part of the casting as the liquid steel envelops them. The external chills are usually made of steel, but in some cases cast iron or copper is used. Both external and internal chills are constructed in numerous shapes and sizes, depending upon the part of the casting to be chilled. The external chills may in many instances be cast to desired shapes so that they will be a part of the mold contour.

Prior to 1930, chills were considered as a solution to the problem of shrinkage cavities. Sections that contained shrinkage cavities were chilled in subsequent castings, generally using internal chills. Only since about 1932 has the chill been looked upon as a method of bringing about directional solidification and of assisting in control of solidification.

External Chills.—The proper application of external chills depends upon a complete understanding of the role that temperature gradients in both metal and mold play in the solidification of a casting. By the use of chills applied to a section, it is possible to solidify that section early in the solidification period of the casting. It is possible to solidify what, under normal conditions, would be hot spots, so that proper feeding of the casting can be controlled and made directional toward the feed risers. The size of the chill employed is of importance, since it is possible to overchill a section and prevent proper directional solidification if too large a chill is used. If the chill is too small, it will fail to accomplish its purpose.

The advantages in favor of the use of external chills over internal chills are

1. Fusion between the chill and the casting is not important in considering the size of the chill; hence the proper use of chills as to size need not be so exact a study.

2. Cleanliness is not so important, since the gases evolved can escape through the sand.

3. In many cases the chills can be used more than once.

External chills have the disadvantage that, for some applications, they must be cast or machined to fit the surface that they are to chill. External chills should never be designed so as to form reentrant angles

of sand at their sides or ends. Such designs may be responsible for sand spalling. The chill should be tapered toward the casting face, for abrupt lines of demarkation between metal chill and sand are positions of cracks due to stress concentration.
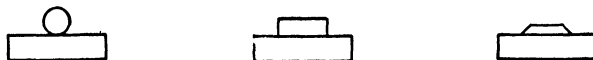
External chills should not be made too long. It is the normal practice in the use of fillet chills to use a chill about 4 or 5 in. long. If a fillet of several feet requires chilling, the 4-in. chills would be interspaced with sand in which is cut a small bracket.

There are but few quantitative studies on the effect of external chills on the solidification of steel. Study has been made of the chilling of a few joining sections.[15] These section junctions were not fed directly with superimposed feed heads but were fed through the arms of the section. The sections studied were therefore typical of inaccessible junctions found in actual castings. The effect of chills upon the defects found in unfed L sections is shown in Fig. 95. The cross-sectional area of the chill, the perimeter of the chill in contact with the casting (noted as effective perimeter), and the cross-sectional area of the defect are given in each case. The section width is 3 in. A chill 1 in. in diameter or a small triangular chill with about the same cross-sectional area is sufficient to cause the area of increased mass at the junction to solidify prior to the

Fig. 95.—Externally chilled L sections. (1) Area of defect 2.8 sq. in.; (2) ½-in.-diameter chill, area 0.20 sq. in., effective perimeter 0.25 in., and area of defect 0.1 sq. in.; (3) 1-in.-diameter chill, area 0.78 sq. in., effective perimeter 0.78 in.; (4) triangular chill, area 0.70 sq. in., and effective perimeter 1.87 in. (*Briggs, Gezelius, and Donaldson.*[15])

solidification of the arms through which the feed metal is supplied.

The effect of chilling unfed T sections is shown in Fig. 96. Chills placed at the radii of a T section tend to move the defect toward the top of the T. A plate chill placed across the top of the T has the opposite effect. If a combination of the two chills is used, the junction will be the initial point of solidification.

An increase in chill diameter at a reentrant angle may not be the proper solution, because in so doing the casting radius is increased. This condition can counteract the effect of the larger chill. The substitution of

triangular chills for circular chills at the casting radius is seldom a solution for increasing the chilling effect at a joining section, since the larger area in contact with the casting and extending down the section arm may cause the arm to solidify prior to the junction and prevent the feeding desired.
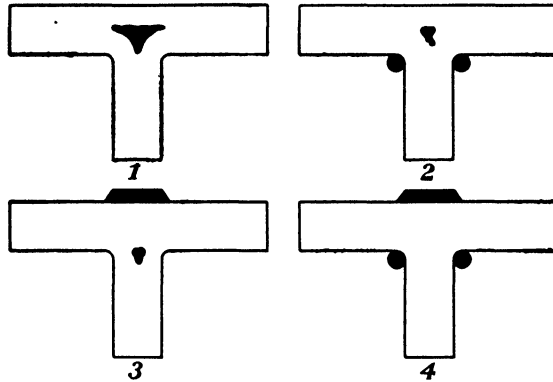
FIG. 96.—Externally chilled T sections.   (1) Area of defect 3.6 sq. in.; (2) 1-in.-diameter chills, area 1.56 sq. in., effective perimeter 1.56 in., and area of defect 0.8 sq. in.; (3) 1-in. plate chill, area 3.75 sq. in., effective perimeter 4.0 in., and area of defect 0.6 sq. in.; (4) 1-in.-diameter chills plus 1-in. chill plate, area 5.31 sq. in., and effective perimeter 5.56 in. (*Briggs, Gezelius, and Donaldson.*[15])
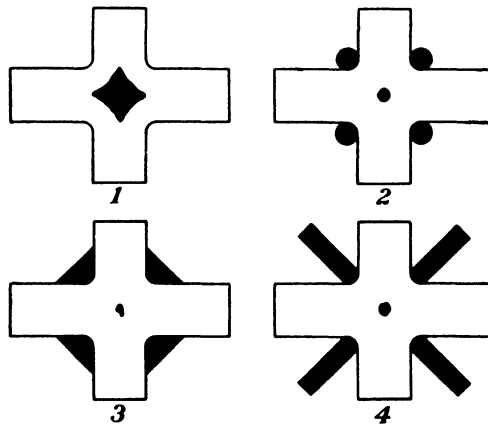
FIG. 97.—Externally chilled X sections.   (1) Area of defect 6.4 sq. in.; (2) 1-in.-diameter chills, area 3.12 sq. in., effective perimeter 3.12 in., and area of defect 0.60 sq. in.; (3) triangular chills, area 10.76 sq. in., effective perimeter 15.2 in., and area of defect 0.2 sq. in.; (4) plate chills, area 15.57 sq. in., effective perimeter 3.12 in., and area of defect 0.5 sq. in. (*Briggs, Gezelius, and Donaldson.*[15])

In T sections, excessive fillets are a liability since large chills are required that act upon the arms of the joining members to solidify the arms prior to the solidification of the junction with resulting shrinkage cavities in the junction area.

The X sections are difficult to chill properly.   In Fig. 97 there are

shown four methods for chilling the X section, none of which produced initial solidification of the junction. In view 3 the insert chilled the arms as well as the junction, causing the arms to solidify prior to the junction. In view 4 an indirect chilling effect on the arms took place. It would seem, as suggested in the section on design, that offsetting the joining arms with some distance between the arms is preferred. These sections could then be chilled as T sections.

Sections in the form of a Y are also difficult to chill successfully. It is difficult to withdraw sufficient heat from the Y section to make it a point of initial solidification, because a chill sufficiently large to be effective at the top of the Y acts as a semi-chill to the arms, resulting in the solidification of the arms prior to the solidification of the junction.

Some general statements regarding the effect of chills on joining sections can be made. The effective perimeter and the cross-sectional area of the chill govern the chilling effect of a particular chill, but the latter is the more important. The heat conductivity of steel is great enough so that the cross-sectional area must be quite large in proportion to the effective perimeter before the efficiency of the chill is impaired. By increasing the effective perimeter of the chill it is possible to bring about solidification in the arms prior to that of the junction.
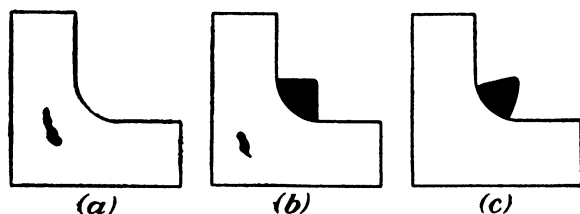
Fig. 98.—Chill treatment of 2-in.-radius L sections: (a) defect obtained when section is cast without the use of chills; (b) area 3.7 sq. in., effectiver perimeter 3.25 in., and area of defect is $\frac{1}{2}$ (a); (c) area 2.9 sq. in. and effective perimeter 2.25 in. (*McKee*.[22])

The importance of the effective perimeter in bringing about the desired chilling effect has also been shown by McKee.[22] In Fig. 98 an L section is illustrated, with 2-in. radius. In the b view the chill covers the entire radius, while in the c view the chill covers approximately three-fourths of the radius. A defect is found when the larger chill is used. If, however, as in the case of the smaller chill, each edge of the chill were one-eighth of the radius removed from the tangent point, the arms would remain open and feed the section junction that is being chilled. A rule proposed[22] for the use of chills at joining sections is that the radii be at least one-third as great as the section thickness but never more than one-half as great.

The chilling of a flat surface offers different possibilities from the chilling of the junction of sections. Quantitative information on the

chilling of flat surfaces is valuable in determining the method of chilling bosses or increased sections that must solidify prior to smaller adjoining sections.

A study has been made of chilling the sides of a rectangular block pattern (5 by 5 by 8 in.) by Troy.[23] Figure 99 illustrates the depth and
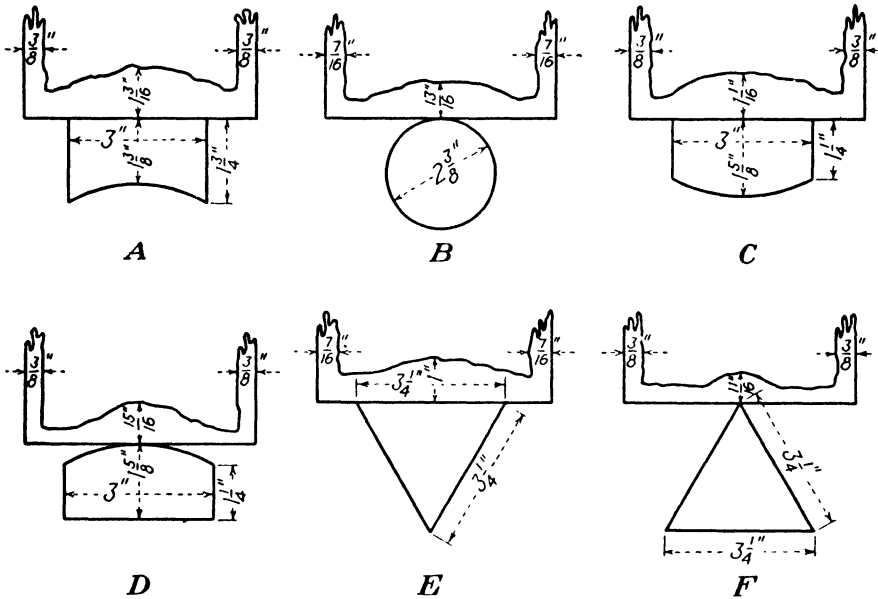
Fig. 99.—Effect of chill shape and contact on the solidification of steel after 65 sec. (*Troy.*[23])

	Effective contact perimeter, in.	Cross-sectional area of chill, sq. in.
A	3.00	4.4
B	0.10	4.4
C	3.00	4.3
D	0.37	4.3
E	3.25	5.2
F	0.02	5.2

extent of chilled area after solidification had proceeded for 65 sec. The chill-effective perimeter and the cross-section area of the chill are given to illustrate their effect on the depth and extent of the chilled area. In order to obtain the maximum heat-extraction rate, the effective perimeter must be about 30 per cent of the perimeter of the chill. It will be noticed from Table XLII that there is but little difference in the maximum depth of the steel during solidification, even though the area of the chill and the effective perimeter are varied.

These studies of chills indicate that

TABLE XLII.—EFFECT OF CONTACT PERIMETER, CHILL CROSS-SECTION AREA, AND CHILL VOLUME ON THE MAXIMUM DEPTH OF STEEL SOLIDIFIED IN 65 SEC. Troy[23]

Contact perimeter, in.	Area of chill perpendicular to mold face, sq. in.	Chill volume, cu. in.	Maximum depth of steel solidified after 65 sec., in.
1.0	5.0	5.0	$\frac{7}{8}$
1.0	1.0	3.0	$\frac{7}{8}$
1.4	3.5	5.0	$1\frac{5}{16}$
1.73	3.0	5.0	1
2.0	2.0	6.0	$1\frac{5}{16}$
2.0	2.5	5.0	1
3.0	3.0	9.0	$\frac{7}{8}$

1. The effective area of heat extraction is greater than the chill-effective perimeter (or contact area).

2. The solidification of a flat surface occurs in a convex shape with its maximum rate at the center of the chill contact.
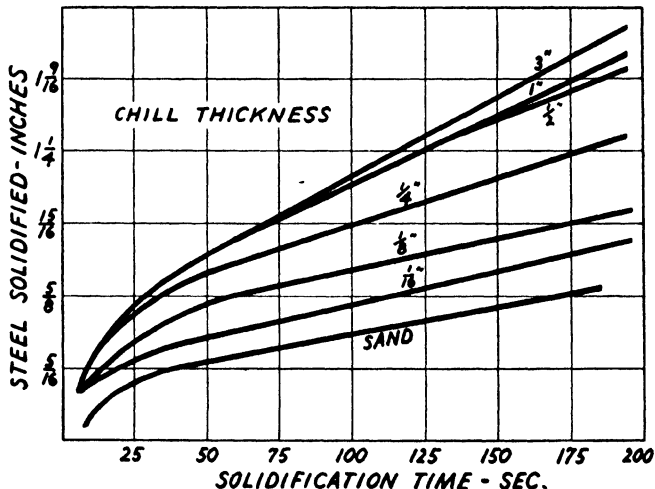
FIG. 100.—Effect of increasing chill thickness with constant-contact area on the rate of skin formation. (*Troy.*[23])

3. If contact is reduced below a desirable minimum, the heat-extraction rate as well as the total heat extraction is reduced, even though excessive chill volume is added.

4. A minimum chill thickness and contact are necessary to establish the maximum solidification rate for any specified time period.

A series of curves (Fig. 100) has been prepared by Troy to show the

rate of heat extraction, using chills of 3- by 3-in. contact and with varying thicknesses. Solidification prior to bleeding ranged from 6 to 195 sec. This figure illustrates that

1. An initial rapid heat extraction is recorded for both sand and chills during the first 6 sec. of solidification.

2. There is a saturation limit for any chill, increasing in time with increased chill thickness.

A saturated chill removes heat at about the same rate as sand.

Information of the type that has been presented is of little value if no attention is given to the fact that the degree of superheat and the amount of steel that passes over the chill will change the chilling effect. A chill can absorb just so much heat, after which the rate of heat transfer of the surrounding sand is the controlling factor, from the standpoint of heat extraction. The chill is of little value if sufficient metal flows over the chill to cause it to reach its saturation limit prior to the onset of casting solidification.

External chills are often given a slight wash of iron oxide or silica flour base, so that there will not be excessive variation in the surface of the casting, since a surface chill will be more effective than sand in producing a smooth casting surface. The type and thickness of the chill coating will alter the effectiveness of the chill.

The development of checks or fine cracks on the casting surface chilled is caused by (1) the use of rough surface chills, (2) massive chills, or (3) allowing the casting surface to adhere to the chill face. To prevent adherence of the cast steel, the external chills must be smooth and protected by a coating.

Internal Chills.—The use of internal chills is regarded with a great deal of suspicion by some purchasers of steel castings. However, their fears are not always warranted, since internal chills can be used properly and effectively. It is true that there have been many cases of the misuse of internal chills. It is known that castings have been produced of a size greater than the furnace melting capacity, by filling the mold with bars and rods and pouring molten steel around them. Such practices are not used today. But practices similar to that just described were responsible for the fact that certain organizations, such as the Navy Department, adopted specifications preventing the use of internal chills.

The primary function of the internal chills is similar to that of external chills, *viz.*, one of permitting a section to complete its solidification prior to an adjoining section so that the liquid feed metal, which compensates for solidification contraction, is derived from the adjoining section.

Internal chills are used primarily in positions of deep recesses or other positions that are difficult to reach as to placement of external chills. They are also used in isolated areas of heavy-metal sections that are

attached to members of much thickness and so located that the foundry-man has no opportunity to feed the heavy portions properly by means of conveniently placed feed reservoir or to chill it by external chills.

Internal chills are used in preference to external chills at locations of machined openings, such as at bolt holes in bosses. Bosses and raised plates are usually areas of increased mass, which, because of their small size, are not profitable to feed directly by attached feed heads. They are potential hot spots, since they are of greater section than the surrounding sections. The bosses are usually cast solid and then drilled for boltholes. An internal chill placed in the position of the hole to be drilled and slightly smaller than the hole diameter can effectively chill the extra metal mass resulting from the boss design, so that the boss solidifies completely prior to that of the adjoining sections. The internal

Fig. 101.—An axle housing steel casting. (*Westover*.[25])

chill is removed in the machining operation, and little thought is given to the possible fusion of the chill with the remainder of the casting.

An illustration of an effective use of internal chills is exhibited in the axle-housing casting of Fig. 101. This steel casting weighs about 600 lb. and is about 44 in. long. If the casting were molded in any position for pouring, it would be found that bosses, or hot spots, were present in almost every conceivable direction. These bosses vary in depth from $1\frac{1}{2}$ to 5 in. In this casting there are 54 hot spots that are machined—either faced, drilled, or drilled and tapped. Internal chills were used in all bosses.

It is suggested by Westover[25] that a good rule to follow is to use a chill with a square dimension equal to one-fourth the diameter of the boss in which it is placed. If the designer has used standard engineering practice, this size chill will be completely removed in the drilling of the boss and will be of sufficient size to produce a solid boss.

The character of the fusion of internal chills into castings has caused some buyers of steel castings to require nondestructive testing at all areas

where internal chills may be placed. Lack of fusion may be responsible for the formation of steam leaks in high-pressure equipment; it also lowers the mechanical properties of a section and possibly results in a centralization point of stresses. Thus the chills must fuse into the cast matrix if a perfectly sound casting is to be obtained. The chill surface must be free from moisture or any other gas-forming material, if blow holes or porous spots are to be prevented.

In green-sand practice the chills can easily be inserted just prior to closing the mold, and moisture condensation on chills and their cleanliness can be controlled. This is not invariably true in dry-sand practice or core-assembly molding, for if the chills must be inserted in the green
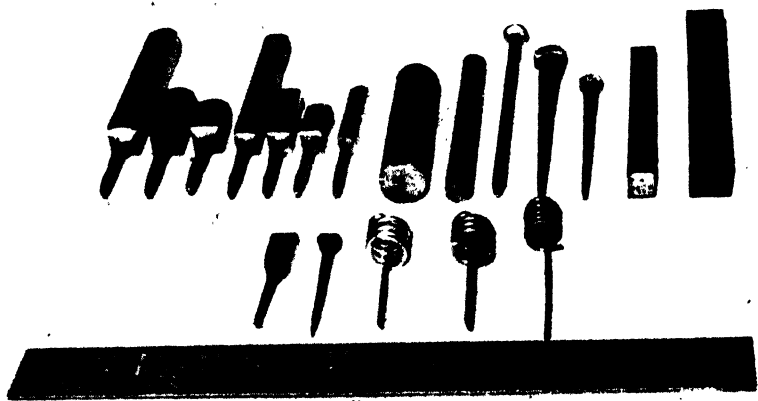
Fig. 102.—Various types of internal chills. (*Westover.*[25])

mold, the fumes from the organic binders will leave a deposit on the chills when the mold is drying. This invariably causes lack of fusion and entrapped gas cavities.

Only a few studies have been made on internal chills, and these are primarily of a qualitative nature.[15,18,24,25,26] These studies have emphasized the importance of the shape or design of the chills, their coating, and their placement and position in the mold. There are a number of types of internal chill being used by steel foundries in the United States. These are plates, bars, coils, various-shaped nails, and specially shaped cylinders. Some of those normally used are illustrated in Fig. 102.

The various sizes of cut nail and horseshoe nail have been used extensively as internal chills, because they are inexpensive and easy to handle. Their main disadvantages are lack of cleanliness and poor shape. The flat top of the nailhead often produces a cavity directly over it. The actual reason for this is not known, but it is believed that the metal, after

rising up the stem of the nail, does not flow smoothly across the top of the head but is cooled and builds up and flows together above the nailhead, entrapping air along the face of the nailhead. The cut nail has the additional bad feature of a reentrant angle at the junction of the stem and the head, which is difficult to fill with molten metal, probably because of entrapped gas pressure.

The coil chill in its many sizes is in common usage in the industry. It was originally produced in this form with the idea that the steel flowing through the coil would key it in place and produce a fusion joint. The

Fig. 103. – Radiograph of a Y section with internal chill. $\times \frac{1}{2}$.

coil chill has a high ratio of surface to mass, and it is believed that this feature is opposite to the fundamentals of effective internal chilling, for mass is required to absorb heat. Surface area should be held to a minimum, for with the increase of surface area the possibilities of lack of fusion increase. Also, too often gas is trapped inside the coil, and the shape of the coil is such as to encourage the holding of dirt. Trapped gas inside coils, which are placed at locations to be drilled, often causes drill breakage or permits gas passages to lead off from the drilled holes.

In many cases the location of the internal chill is shown clearly on the radiograph by lack of fusion, dirt, or gas around the chill, such as illustrated in Fig. 103. In this illustration the lack of fusion is not due to overchilling, since the type of defect is the same whether one coil chill or four are used. The use of solid internal chills presents the same difficul-

ties, and it appears that, for the same volume of chilling material, the one with the least surface-contact area would give the best radiographic results from the standpoint of fusion.

Some experimental work by Lanzendorfer[24] tends to show that best fusion results are obtained when the chill weighs between 2 and 5 per cent of the total weight (chill plus steel) of the section to be chilled. Below this value the fusion was good, but the chill was insufficient to prevent hot-spot formation and the subsequent shrinkage cavity.

In experimenting with internal chills, Westover[25] found that when nails were used as chills with their points up, the blows, or entrapped gas, were less objectionable than where they were supported with surfaces, and particularly their top sides, in horizontal planes. Further experiments disclosed conclusively that by the use of chills specially shaped and positioned so that there were no horizontal surfaces, the blows or "chill kicks" were eliminated. A solid chill of square cross section with a pyramid top was adopted as being the ideal chill.

It is obvious that when a square solid chill with a pyramid point is placed in the mold with its longitudinal center in a horizontal or in a vertical position, the chill can be so located that all its surfaces will incline upward so that there will be no horizontal top surface.

Internal chills should be devoid of moisture, grease, or gas-forming materials, devoid of refractory material, and presumably free from rust. To ensure freedom from rust and to increase the ease with which the chill fuses into the casting, it has been the practice to coat the chills. An investigation of various chill coatings by one of the technical and operating groups of the Steel Founders' Society concluded that a copper plating produced the best all-round coating for internal chills. The subject of coatings will be discussed in more detail under Chaplets (page 239).

Sections cut from castings containing internal chills will show equally as good mechanical properties as similar sections containing no internal chills, providing the chills are well fused into the surrounding cast metal and they are of a composition similar to the cast metal. If chills are to remain in alloy castings, the chills should be produced from steel that will give mechanical properties equivalent to those of the casting after the prescribed heat treatments. If the chills are to be removed from the finished casting by machining, then the usual mild steel chills can be used.

The procedure for the placement of chills in a mold varies considerably from one foundry to another. It has come to be acceptable practice to place small aluminum or brass buttons or tacks on the pattern, to show the location of internal chills. In most cases green-sand molds containing internal chills are not closed until just prior to pouring. In dry-sand practice the chills are usually inserted in the mold prior to drying. Con-

siderable trouble due to blow holes and lack of fusion has been caused by such a practice. It is, however, difficult to insert chills into hard, dried molds. Further work is desired on the process control of chills in molds requiring drying.

No tables have been prepared or formulas given for the calculation of the chill volume for the chilling of varying sections. Experimentation of the trial-and-error type on individual castings is necessary to establish proper chilling requirements.

Since chaplets are a form of internal chills, it is suggested that the following topic also be reviewed as a part of the study of internal chills.

Chaplets.—A chaplet is a metal construction to keep cores in position while the mold is being poured with molten metal (Fig. 104). There are a great many designs and sizes of chaplets regularly used in the steel foundry.

Certain purchasers of steel castings look with even more disfavor upon chaplets present in steel castings than upon internal chills. Since a chaplet extends from one cast surface to another through the cast section, it is possible that "weeping" will take place under hydraulic testing or comparable service conditions owing to lack of fusion between the chaplet and the cast metal. Also, areas near chaplets often show porosity, and in some cases hot tears.

FIG. 104.—Chaplets to hold cores in place. (*Stimpson, Gray, and Grennan, "Foundry Work."*)

There has not been much information available regarding the successful applications of chaplets. Generally a foundry will carry several sizes and designs as stock, and the molder gives more consideration to the dimensional characteristics than to weight of construction. The proper application of chaplets is the function of the planning committee of the shop, and as much care should be given to their selection as would be given to chills or gates and risers. It has been said by certain foundrymen that the proper application of chaplets is to eliminate their use as far as possible, since they are always a potential source of casting defects. Usually ways and means can be adopted to eliminate the use of chaplets, but these methods are more costly in mold-assembling time. However, on pressure castings the extra time and costs are worth the effort.

It has been pointed out that there are many variations of chaplet design. However, some experimental work by Taylor and Rominski[26]

shows that there should be no reentrant corners where gases could be trapped. Also, the screw threads on the stem are of no particular aid in the fusion of the chaplet with the cast metal.

Chaplets are ordinarily made of low-carbon steel that can easily be stamped, rolled, machined, and assembled. Some high-alloy chaplets are available, such as Monel-metal chaplets, but these are not often used in carbon and low-alloy cast steels.

A core surrounded by molten metal follows the laws of buoyancy; *i.e.*, it is buoyed up by a force equal to the weight of the liquid displaced.
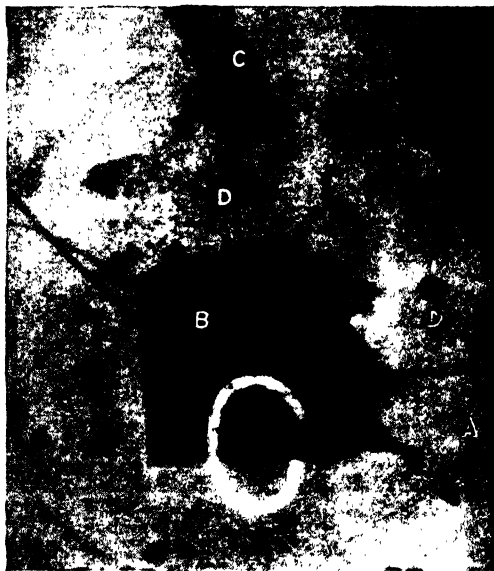
FIG. 105.—Radiograph of section of steel casting showing: (*A*) hot tear surrounding dark square; (*B*) dark square is surface depression caused by chipping out square wedge used to build up chaplet; (*C*) sand inclusion; and (*D*) areas of porosity. $\frac{1}{2}$ reduction. (*Briggs and Gezelius.*[31])

Thus the lift of a sand core in molten steel is 3.9 times its weight. Chaplets supporting cores must support only the weight of the core. They, however, must be large enough in section to withstand the flow of metal past them, but not so large that they will not fuse readily with the cast metal. Chaplets set to hold cores from lifting must withstand a safe load of about four times the weight of the core.

It is common practice for foundrymen to consider strength and rigidity as the primary prerequisites of a chaplet. As a result, many chaplets are used that are much too large, considering the core-weight requirements. Such chaplets do not fuse so easily as the smaller ones at the normal pouring temperatures employed. The ratio of cold chaplet volume to the heat content of the contacting molten metal may be such as to preclude

any possibility of fusion. If chaplets are too large, they act as points of initial solidification such as external chills and, since directional solidification is not planned, may result in hot-tear defects such as shown in the radiograph of Fig. 105.

Many different coatings have been tried by foundries with respect to the preparation of chaplet surfaces. A coating is usually advisable, to prevent rusting of the low-carbon steel. Care also should be given so that water films or oil films are not present on the chaplet at the time that the mold is poured.

Chaplets and internal chills have been coated by dipping, electroplating, and impregnating. In the past most chaplets and chills have been dipped into molten tin. A thin layer of the tin adheres to the steel chaplet. Studies[26] on electroplated coatings of nickel, copper, cadmium, and silver, in thicknesses from 0.0005 to 0.01 in., have been made. Chaplets inpregnated with silicon have also been studied. The results of the test work by Taylor and Rominski[26] on chaplet coatings showed that silicon-impregnated chaplets fuse readily, but that they are difficult to keep free from moisture and dirt because of their very rough, flaky surface. Tin-dipped chaplets gave good results at times, but the results could not be counted upon for consideration. Copper- and nickel-plated chaplets proved very satisfactory, but the most consistent results were obtained with silver-plated chaplets. Aluminum-sprayed chaplets were also thought to have merit.

A black iron oxide coating for chills and chaplets has been reported by the Naval Research Laboratory as giving good results. This treatment is not strictly a coating, in that it is not dimensionally additive to the originally untreated chaplet or chill, but there is an increase in thickness, though slight. This treatment results in the formation of a coating of Fe_3O_4, of from 0.0001 to 0.00015 in. in depth, produced by immersion in a salt bath high in sodium hydroxide.

Chaplets should be sandblasted prior to coating, in order to be assured that the surface is properly cleaned. Surface preparation prior to coating appears to be more important than the type of coating, providing the coating itself does not volatilize or segregate to cause gassiness or poor properties in the casting.

It has come to be accepted practice in many steel foundries today to chip out the head portion of all chaplets and reweld the section so that there can be no possibility of "weeping" along unfused chaplet faces when hydrostatic testing is applied.

CENTRIFUGAL CASTING

The term "centrifugal casting" is an all-embracing one and refers to any casting that has been rotated during the time it solidified. There

are, however, three types of centrifugal steel casting. These are

1. True centrifugal castings (Fig. 106)
2. Semicentrifugal or profiled centrifugal castings (Fig. 107)
3. Pressure or centrifuged castings (Fig. 108)

True Centrifugal Castings.—True centrifugal castings are obtained by pouring steel in a mold rotating around a horizontal or vertical spindle. The mold may be either a metal die or a sand-lined metal jacket. The

FIG. 106.—Weld-neck casting, true centrifugal type. (*Donoho.*[32])
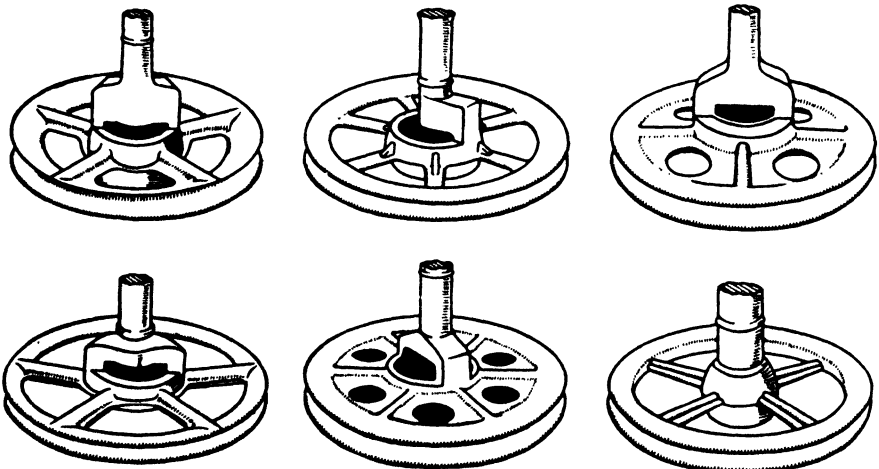
FIG. 107.—Sheave-wheel castings, semicentrifugal type. (*Johnson.*[33])

steel solidifies from the die-casting interface towards the center of the casting. There are no gates or risers, and yields approaching 100 per cent may be obtained. The casting may or may not have a cylindrical hole through its center, depending upon whether or not a measured amount of molten steel is poured. The central region of the casting is usually finished by machining.

Semicentrifugal or Profiled Centrifugal Castings.—The molds of these castings may be either dies, sand, or core-assembly molds where the inner surfaces of the castings are not formed by centrifugal force alone but are formed by cores, with low speeds employed. In this case solidification occurs in different directions with the result that shrinkage cavities may appear unless feed heads are provided. Gates of various types may be used. Some gates may serve as a feed head, and others merely as a pouring basin.

Pressure Castings.—This type of casting refers to the use of centrifugal action generating a high pressure to force the molten steel into the mold. This method is used for unsymmetrical castings, which cannot be spun about their own axes. The mold is arranged so that steel is poured into a gate at the center of rotation. Ingates distribute the steel to mold cavities arranged around the central down gate.

Advantages of Centrifugal Casting.—The primary reasons for casting steel centrifugally are (1) to produce castings more economically, and (2) to cast a part that can be more satisfactorily cast centrifugally than statically. A steel casting of a certain design, such as a wheel or a cylinder, when cast statically may result in a yield of from 50 to 65 per

Fig. 108.—Centrifugal steel castings, pressure type. (*Courtesy of Duncan Foundry and Machine Works.*)

cent. The same casting may be centrifugally cast with a yield of 65 to 85 per cent. The increased yield usually results in reduced cleaning-room costs.

Circular or tubular castings are of ideal design for centrifugal casting. When these objects are made statically, there is a tendency for the development of shrinkage cavities in areas inaccessible to feed heads. This is especially true if high yields are required. In this case the centrifugally produced casting would be of a higher quality. Centrifugal steel castings that solidify progressively from both outside and inside surfaces may also be of inferior quality unless precautions are taken to feed the last portion of the casting to solidify. The metal solidifies from the outside surface only in the true type of centrifugal castings, and there is no possibility of shrinkage cavities' occurring.

When a mold is spun about a horizontal axis, such as a true centrifugal casting, the interior cavity formed by the molten metal is perfectly cylindrical and is exactly determined by the mold dimensions and the volume of the metal poured.   Solidification takes place generally only from one face, *i.e.*, from the mold-metal interface toward the center of rotation. Under certain circumstances, depending upon the thickness of the metal to be solidified and the diameter of the mold opening, the steel may solidify from the inner face, causing center-line shrinkage.   This condition is not desirable.

A mold containing molten steel spinning about any axis other than horizontal will produce, according to Donoho,[29] an interior cavity of a paraboloid shape, the dimension of which depends upon

1. The spinning speed
2. The maximum diameter at the top of the cavity
3. The angle of inclination of the axis of revolution

When these factors are known, the cavity, except as modified by contraction cavities, can be calculated.

Vertically rotated castings are usually gated on the axis of rotation. The gate may or may not act as a feed metal reservoir.   In some cases wheel-shaped castings with central cavities are permitted to solidify without attached feed heads.   A section through the paraboloidal cavity at the axis of the casting carrying the shrinkage and nonmetallic inclusions is bored out in producing the finished product.

Fundamentally, there is no difference in the solidification of casting when the centrifugal casting process is used.   The same laws are effective, considerable use being made of directional solidification. The only difference is that directional solidification is applied so that advantage is taken of the definite pressure exerted by centrifugal force from the center of rotation to the periphery of the casting during its rotation.

All centrifugal casting is, in effect, pressure casting, for the molten metal is forced against the mold wall under relatively high pressure as compared with the usual gravity head in static casting.   This assists toward a better filling of the mold cavity with metal, and greater availability of metal for feeding purposes.   Pressure as applied by centrifugal force differs from that applied by other pressure casting methods, since it is selective in its action, the heavier components being thrown outward with greater force than the lighter particles.   This helps to eliminate light nonmetallics by flotation inward toward the axis of the rotating molds.   Slag, bits of furnace or ladle lining, or sand included in the steel will thus be largely eliminated.

The centrifugal force displacing metal from the center of rotation to the periphery varies directly as the diameter and as the square of the

speed. A simplified formula for centrifugal force is

$$CF = 0.000341 Wrn^2$$

where W = weight in pounds, r = radius of rotation in feet, and n = number of revolutions per minute.

In a statically poured casting the metal is under atmospheric pressure of 14.7 lb. per sq. in., and usually an additional 1 to 2 lb. per in. because of hydrostatic pressure. To these values can be added the pressure of centrifugal force in castings solidifying under centrifugal action.

It has been previously pointed out that during the process of solidification dendrites tend to grow out into the liquid metal, forming a mass of interpenetrating spines that often "landlock" the metal between them. These small pools of liquid metal, when they solidify, are rather isolated, cut off from a supply of liquid metal to replace the volume contraction due to solidification, thereby resulting in micro- or macro-cavities. Under centrifugal action these voids near the solidifying face are prevented from forming since the force is greater than the strength of the solidifying steel. This does not mean that there are no shrinkage cavities in centrifugal castings. Any casting that solidifies from more than one surface, the last point of solidification of which is not fed by reserve metal, will exhibit shrinkage cavities regardless of whether it is cast statically or centrifugally.

High values of centrifugal force are not necessary; *e.g.*, Donoho[28] states that a general rule to follow for sand-lined molds is to use a spinning speed that will give a centrifugal force of about 75 times gravity. A simple formula derived from the centrifugal force formulas to give the spinning speed for a centrifugal force of 75 lb. per pound of metal is

$$n = \frac{1,675}{\sqrt{R}}$$

where n = revolutions per minute and R = radius in inches.

Spinning speeds at or near the favorable revolutions per minute, as calculated by this formula, have been found generally best, particularly for the initial spinning speed during pouring. Excessively high spinning speeds will tend to cause longitudinal hot tears in the casting by reason of the circumferential tension developed by the centrifugal force as the casting solidifies. Spinning speeds for castings made in metal molds may be somewhat lower than the calculated value, since the chilling effect of the mold causes a greater rate of solidification than that in sand molds.

It is apparent from general observations on the method that the centrifugal force makes it possible to feed a greater length of section from a pool of liquid metal than when this force is absent. It also seems that under very definite limits of design it is possible to feed heavier sections

through slightly lighter sections, taking advantage of metal and mold temperature gradients and centrifugal force. An example of such a condition is in centrifugally cast gears, with the exception that the solidity of the rim depends on the relation of the cross-sectional area of the feeding channels (spokes) to the cross-sectional area of the rim. Studies by Wright and Caine[29] show that the cross-sectional area of the channels, whether they be spokes or enlargements of the web, must be equal to the cross-sectional area of the rim in order to make a radiographically sound gear regardless of the size of gear or the speed of rotation. However, this ratio can be increased to 1.2 to 1 (rim to spoke) before the shrinkage area extends from the spoke into the rim.

The length of the channels (spokes) is another variable listed by Wright and Caine[29] as affecting feeding in centrifugal casting. With distances between the outside diameter of the hub and the inside diameter of the rim less than 4 in., no trouble due to lack of proper feeding will be encountered with feed channels of uniform cross section. With greater length it is necessary to pad or taper these channels in order to obtain directional solidification. They suggest a taper of at least ¼ in. per ft.

It has been stated from time to time that centrifugal castings are of better quality, are denser, and have mechanical properties in excess of those of statically poured castings. If a true centrifugal casting were being compared with a poorly fed statically poured casting, the statement is correct, since the latter would most probably contain shrinkage cavities or areas of center-line weakness. Test specimens taken in this zone would not exhibit the density or the mechanical properties of test specimens taken from the true centrifugal casting. If a well-fed statically poured steel casting is compared with a true centrifugal steel casting of similar design made in a metal die mold, there will be no difference in the density of the two castings.

In tensile specimens taken from the heat-treated castings in similar locations, it will be found that near the die face higher mechanical properties will be recorded than can be obtained from a similar area in a sand-mold, statically poured casting. This is because of the finer structure produced by the chilling action of the metal die. However, if a well-fed, statically poured steel casting is compared with a similarly designed profiled or semicentrifugal casting consisting of a sand mold and core inserts, no difference between the density or the mechanical properties will be found, since rates of cooling and effect of mass conditions are similar.

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CHAPTER VII

GATES AND RISERS

GATES

Considerable importance is attached to the manner in which molten steel should enter a mold. Steel foundrymen have long been aware that a successfully made casting must be properly gated. The purpose of the gate system is, primarily, to furnish a means of allowing the steel to enter the mold. Although this is the main purpose of a gate, there are other functions that a gate can and must perform.

A study of the operation of pouring a mold will show that the aims are as follows: (1) to fill the mold cavity completely, (2) to avoid the deterioration of the mold by the liquid metal, (3) to establish proper temperature gradients in the mold so that the gate will not be responsible for the occurrence of contraction cavities in the casting.

Measures to be taken to accomplish the desired aims include the control of: (1) the filling method, *i.e.*, the manner by which the liquid metal is transferred from the ladle to the mold; (2) the type of runner that controls the disposition of the gates carrying the liquid metal into the main mold cavity; (3) the position, dimensions, and number of gates entering the mold cavity; (4) the rate of pouring; (5) the position of the mold during casting; and (6) the temperature and fluidity of the metal. Pouring is an important phase in making a casting and often has a preponderant influence on its successful production.

The gate system, besides filling the casting with molten steel, also may function as (1) a venting system for the removal of gases from the mold, and (2) a riser, to furnish liquid metal to the casting to compensate for volume losses due to liquid and solidifying contraction.

Nomenclature.—For years, efforts have been made to standardize the terms used to explain gating methods. Colloquialisms and the indiscriminate use of some terms have added to the confusion. Even the technical literature is not immune to these difficulties of interpretation from the interchangeable use of terms. A subcommittee of the Committee on Nomenclature of the American Foundrymen's Association reported on gating terms in 1940. The report is based on the gating needs of the entire foundry industry, and certain gates discussed are not used in the gating of steel castings.

The terms recommended were selected as those most commonly used in practice, and also the most descriptive. For example, the term "sprue" was selected as a proper term in preference to "downgate" and others, inasmuch as the latter term would be so similar to the recom-
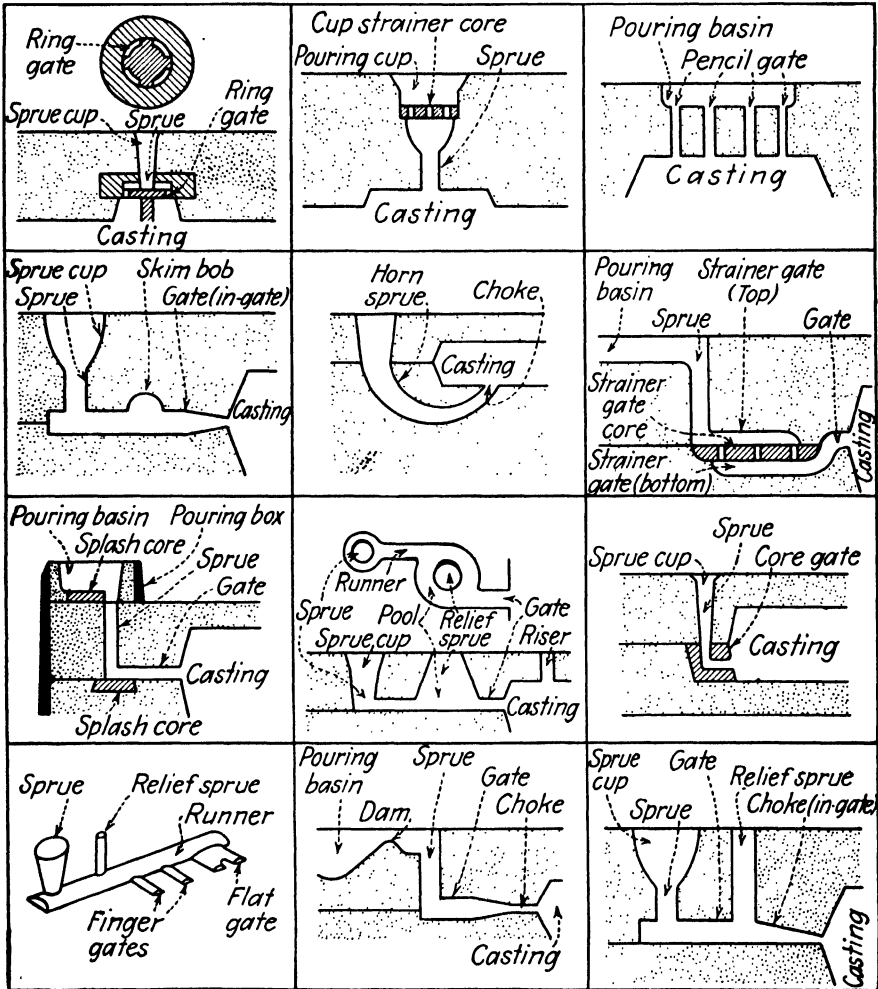
FIG. 109.—Nomenclature of gate systems.[37]

mended term "ingate." Illustrations of nomenclature, as recommended for various types of gating arrangement, are shown in Fig. 109. The illustrations point out the proper nomenclature, and they do not necessarily illustrate the types of gate used for steel castings. The most common gate for steel castings is shown in Fig. 110.

Gate Types.—Gating of steel castings is a subject of considerable controversy. Opinions have been formed and re-formed, and one foundryman's experience has clashed with another's until one begins to realize that there can be no question of orthodoxy but rather a question of sensible application based upon fundamental principles, and of recording of past experiences with different types of casting from the standpoint both of fundamentals and of past experiences.

Certain factors should be considered in planning the type of gate to be used. These factors are (1) the type of gate best suited and most convenient to use on the particular casting that requires gating; (2) the size and shape of the gate system, since these items govern the speed at which
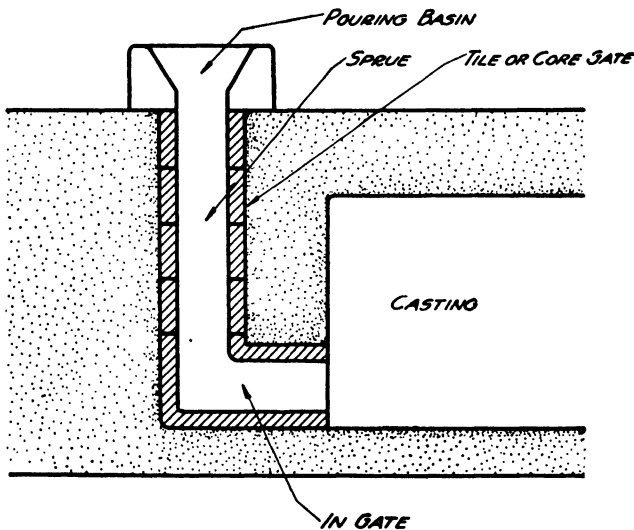
FIG. 110.—Typical bottom gate for steel castings.

the mold is filled; and (3) the possible need of slag or inclusion traps, depending upon the decisions made in regard to points 1 and 2, in order to ensure clean metal's entering the mold.

The literature on the subject of gating shows a preponderance of the use of bottom gates for steel castings. This gate system is of two types: (1) the typical bottom gate, as shown in Fig. 110, and (2) the horn-gate construction, examples of which are shown in Fig. 111.

The typical horn gate has a disadvantage in producing a fountain effect within the casting, causing undue mold erosion and the entrapping of air. In general, horn gates should not be used to gate steel castings unless the inverted horn gate is used, such as shown in Fig. 111, where the large end of the horn enters the mold cavity. In general, a safe rule to follow is to permit horn gates to enter only the risers or blind risers and not allow them to enter the mold cavity proper.

Bottom Gates.—The consensus of opinion[1,2,10,14] is to the effect that whenever possible a mold should be gated at the bottom. This presumably allows the easy flow of the metal into the mold, thereby reducing the erosion of the mold surface by the molten steel. It is also stated that
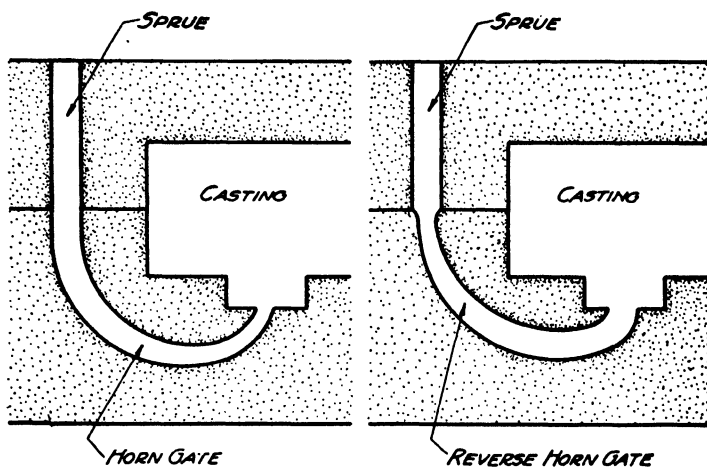
FIG. 111.—Horn-gate and reverse-horn-gate construction.
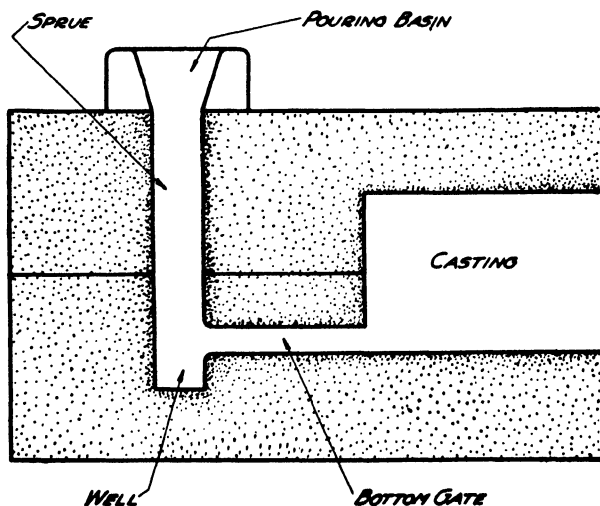
FIG. 112.—Bottom-gate construction employing the use of a well.

the use of the bottom gate prevents cavities in the casting, since there is less danger of trapping air in the mold. By the use of bottom gating it is easier to design the gating system so that the steel entering the mold will not impinge on inserted cores or projecting parts of the mold.

The bottom gate is often constructed so that a well is included as a part of the gate system, as in Fig. 112. The purpose of this well is to act as a cushion for the metal dropping down the sprue, so that the metal will not erode the gate at a position of radical change in the direction of flow.

When metal is dropped some distance, a change is made in the direction of flow of the steel to prevent the erosion of the sprue and the ingate. The steel is allowed to drop to the parting, where it flows horizontally or at an angle to the horizontal, before continuing its further drop. These arrangements are shown in Fig. 113. The sprue and the ingate are in
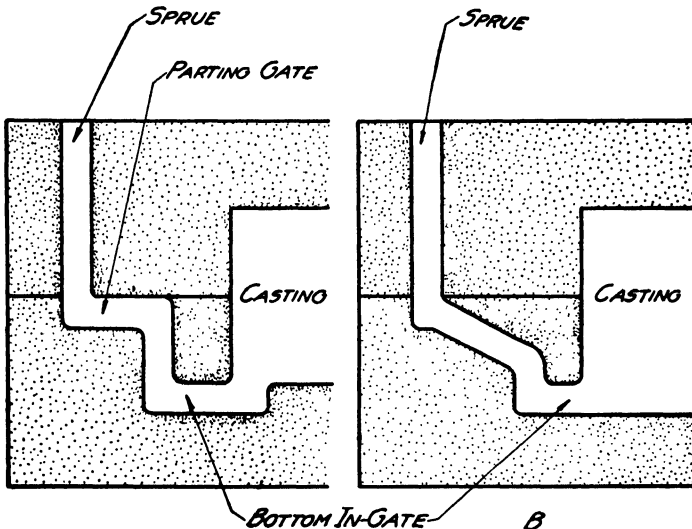
FIG. 113.—Bottom gates, showing change in direction to prevent mold erosion.

most cases prepared in cores or tile. The literature reports many variations of gates, such as shown in the examples in Fig. 113.

Parting Gates.—There appear to be numerous types of gate that enter the casting at the parting joint. The one outstanding rule that has been followed in conjunction with these gates is that whenever possible the metal from the ingate should enter the mold directly under the riser.

Occasionally a whirl gate is used in conjunction with an ingate on the parting joint. Examples of these gates are shown in Figs. 114 and 115. The purpose of these gates is to collect all slag, eroded sand, etc., and trap it in a gate or riser that comes to the surface, allowing clean metal to enter the casting. The gate system, as shown in Fig. 115, may or may not be in a line. In fact, in most cases the gates are bunched so that they will not take up too much room in the flask.

Another type of whirl or swirl gate is used in conjunction with large gears and rolls (see Fig. 116). The purpose of this gate is to require the

metal to flow one way around the gear. This prevents a cold shut or an oxide lap from forming, as would be the case if the metal were allowed to flow both ways at once and to meet in the casting on the side opposite the gate. In this gate system it is important that the ingate enter the casting on a tangent. If this rule is observed, the metal will not run in the cast-
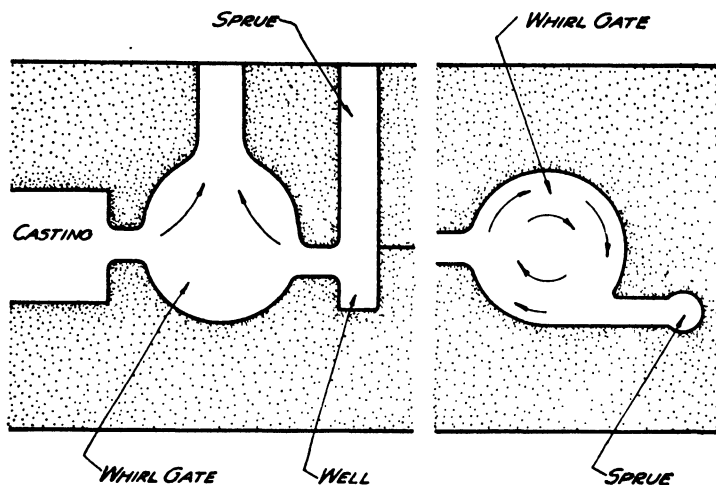
FIG. 114.—Sphere whirl gate.
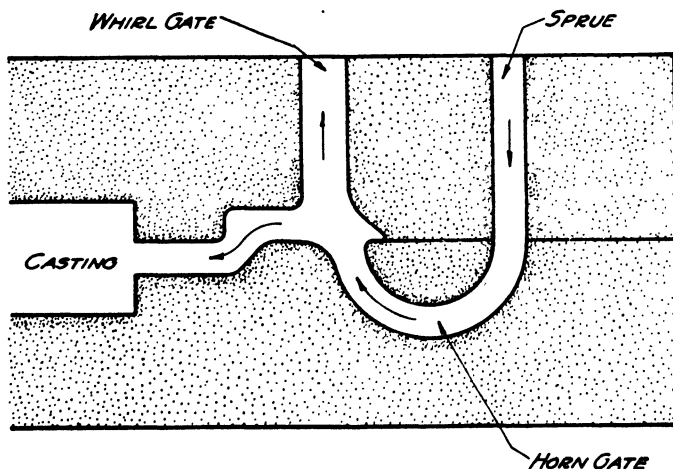
FIG. 115.—Horn- and whirl-gate construction.

ing section behind the gate to any great degree. It is claimed by those who use this swirl type of gate that because the metal swirls in the mold, all sand, slag, etc., will be carried along on the top of the stream and eventually be deposited in the risers, especially if the riser covers the entire casting, as it does in rolls and certain gear castings. The ingate is constructed in two ways. In one case, such as in the production of

rolls, a nozzle effect is used to assist in the swirling motion. In cast tooth gears, the wide orifice, as shown in Fig. 116, is commonly used.

Top Gates.—Gates for the top gating of steel castings are varied in their appearance and show types similar to those used in the pouring of
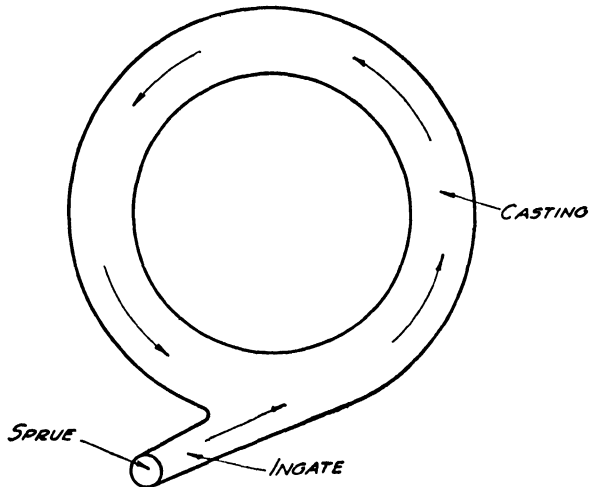
Fig. 116.—Tangent gate for large gears.
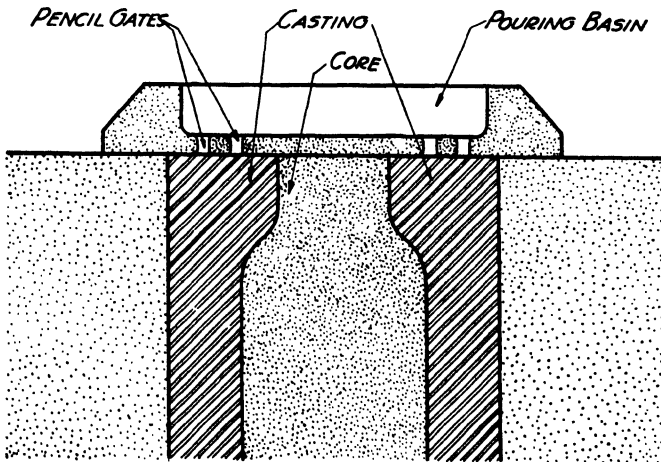
Fig. 117.—Pencil gates for top gating of steel castings.

cast iron where the metal is poured directly down the riser. In other cases the metal will enter at the base of the riser.

Most foundrymen will agree that it is not profitable to employ top-pouring methods in the majority of cases. However, when it is possible to run a casting from the top, far better results are obtained when a number of gates are used instead of one common gate.

A shower gate similar to those used in cast iron is occasionally employed for steel. A gate of this type is shown in Fig. 117. Strainer cores can be inserted in any type of gate. These cores are usually placed at the bottom of the pouring cup, at the entrance of the sprue. The
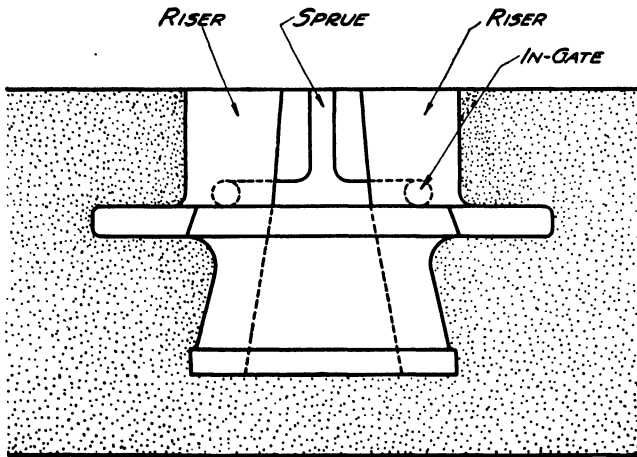
FIG. 118.—Gating into the base of risers. (*Johnson.*[34])
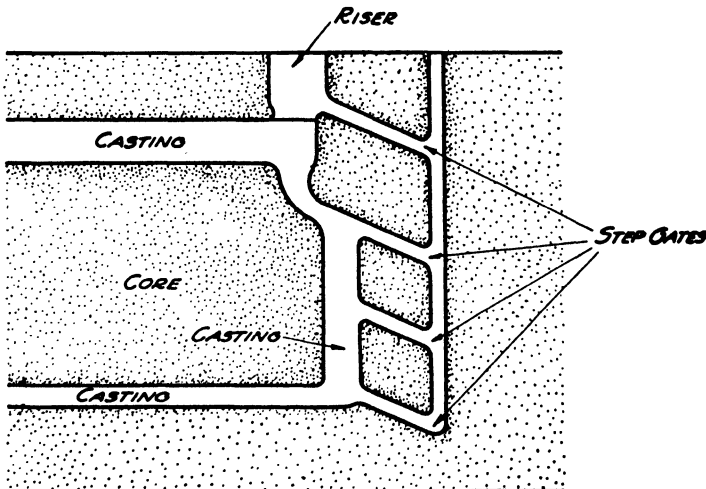
FIG. 119.—Step-gate construction.

method of pouring directly down the riser is frequently employed, particularly in the running of small castings where only a single riser is used and it is adaptable for a gate. Gates that enter at the base of risers are used to a considerable degree for the running of steel castings. An example of such a gate is shown in Fig. 118.

Step Gates.—A combination of the bottom gate, parting gate, and top gate is found in the step gate. This type of gate, an example of which is shown in Fig. 119, combines the features of bottom pouring with those of supplying hot metal to the riser system. Those who use this type of gate point out that when metal rises to the next step, the lower ingate becomes practically inactive.

Rule-of-thumb Methods of Gating.—Gates for steel castings may vary from a fraction of a square inch in cross section to as high as 7 or 8 in. in diameter. Gates may be single, double, or multiple. In shape they may be Y, fan, horn, straight, or step. They usually are so arranged that they will not cut, wash, or trap air.

Gates must not be the origin for casting defects, such as hot tears or contraction cavities. Some attention must be given to the locating of gates with regard to the feasibility of their removal with a minimum of expense. A number of practical rules regarding gates have been postulated from time to time. These may be listed as follows:

1. A round gate will pass more metal than will a rectangular gate of equal cross-sectional area, since it has a lower surface friction.

2. To make a whirl gate work, the diameter of the whirl gate should be larger than the spruc, and the whirl gate should have a larger cross-sectional area than that of the ingate.

3. Whenever possible, all gates should be formed by the use of patterns that are a part of the main casting pattern, rather than cut gates, because they withstand better the metal-erosion action.

4. Ingates should be so directed that the incoming flow of metal takes place along channels in the mold; in this way the channel fills rapidly, allowing the metal to overflow onto the adjoining surfaces.

5. Metal should enter the mold at as many points as possible in order to avoid hot spots.

6. To run high castings, the end of the mold where the ingate is located should be lowered so that the metal is compelled to run up a slight incline. By doing this the metal face travels in a body.

7. The area of the cross section of the ingate should be kept smaller than that of the casting; otherwise, when the gate solidifies, it will pull a piece out of the casting.

8. The section thickness of a casting must be considered in relation to the pouring rate (rise of metal in mold per second); it is common foundry practice to ensure a high pouring rate for light sections.

9. The ingate at the entrance to the mold should not be choked. This allows the metal to enter the mold with increased velocity, and the mold may not be able to withstand this eroding force.

10. Choking of the ingate to assist in sprue removal is a proper procedure, providing that the total cross-sectional area of the ingate is relative to the cross-sectional area of the sprue.

Statements 9 and 10 seem to be somewhat at variance. Postulate 9, however, refers to the use of a single ingate whereas postulate 10 agrees that choking is a poor procedure, unless there are a number of ingates so that the metal will not increase in velocity when it enters the mold.

It is well to mention here a point concerning the mode of solidification of the two gates. In the case of the enlarged ingate, as is shown in Fig. 120, sketch 1, the sprue B will solidify before point A, and the ingate will draw metal from the casting with the result that a contraction cavity may be formed in the casting. With the reduced ingate, as in Fig. 120, the point A will solidify first and will draw metal from B and not from the casting. Also, an important feature in favor of a number of ingates is that local heating of the mold at the ingate area is reduced proportionately to the number of ingates employed.
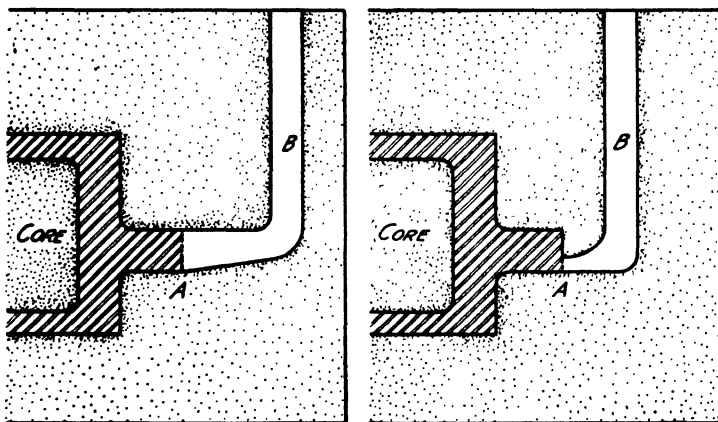
Fig. 120.—Left sketch—enlarged ingate; right sketch—reduced ingate.

Reasons for the Use of Different Positions of Gates.—In general, the parting-line gate is used, since it is the easiest to construct, especially if the gate is cut by the molder. Another feature of this gate is that it usually is possible to put the gate directly under a riser, since the gate definitely causes a hot spot, and all hot spots must be fed. The best way to compensate for a hot spot is to establish a riser directly over the hot-spot area.

It should be remembered that all the metal entering the casting enters through the gate. The mold in the vicinity of the gate becomes heated to such an extent that it retards the cooling of the steel, and the rate of skin formation at this part of the mold is not so great as at other parts of the mold. The result of this condition is that a hot spot is formed. Hot spots are the last points of solidification in a casting, and unless there are means of supplying this portion of the casting with an adequate reservoir

of molten metal to compensate for the solidifying contraction that takes place at this point, a cavity will be formed.

Gates at the parting joint, placed directly under a riser, are usually free from the contraction cavity that has a tendency to form when unfed hot spots are present. A disadvantage of the parting-joint sprue is that it is necessary for the molten steel to drop in the mold in order to fill the drag portion of the mold. Such a drop often is responsible for mold erosion.

In order to keep mold erosion to a minimum and to allow the metal to rise quietly around cores, molds are bottom-gated. This distinct advan-
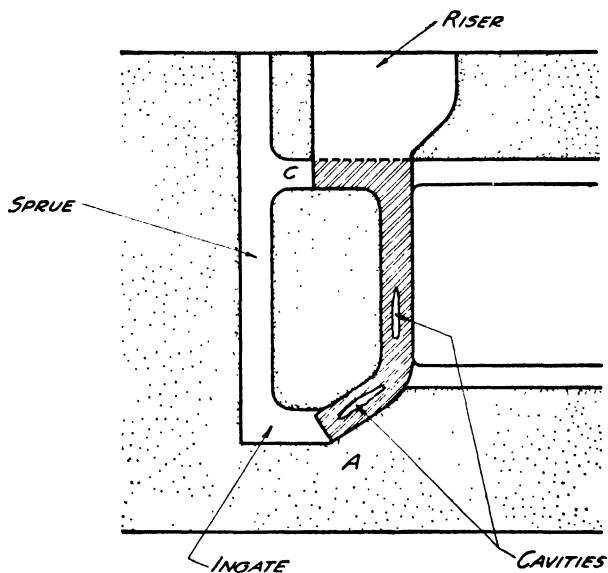
Fig. 121.—Cavities formed by hot spots resulting from bottom gating.

tage of mold gating is offset by an equally important disadvantage, *viz.*, the formation of localized hot spots. These localized hot spots are more serious in character than those usually formed by gating at the parting joint, since, in bottom gating, usually no riser is present to prevent cavity formation due to the presence of the hot spot. An example of this condition is shown in Fig. 121, where bottom pouring is responsible for the formation of the defect at point *A*, owing to the setting up of undesirable temperature gradients. This cavity is not prevented by the riser at *B* regardless of its size and of the fact that hot metal enters the riser as provided for by the gate at *C*.

Another disadvantage of bottom pouring is that, as the metal rises in the mold, it gives off heat to the mold so that the metal that finally goes into the risers is relatively cold metal; *i.e.*, it is considerably colder than

the metal adjacent to the gate.   Such a set of circumstances is just the reverse of that desired in a casting.   The hottest metal should be situated in the riser, and the point in the casting farthest removed from the riser should be the final resting place of the coldest metal.

Thus bottom gating of a casting inevitably produces an adverse temperature gradient in the metal, the top of the casting being coolest and the bottom being hottest at the moment the mold proper is filled.   The degree of this temperature difference and the angle of the temperature gradient are related to the pouring rate, rise of metal in the mold per second, and to the heat conductivity of the mold.   A slow pouring rate will produce a poor temperature gradient, as compared with that produced by fast pour-
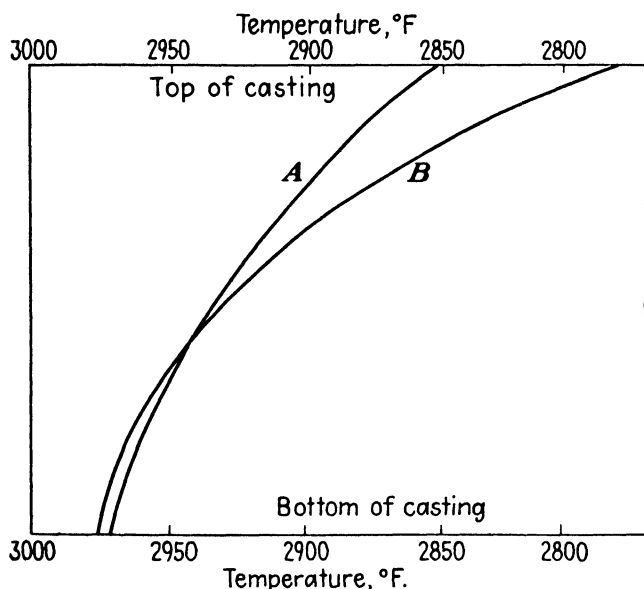
Fig. 122.—Temperature gradients in metal in bottom-gate systems.   (*Bally.*[16])

ing when the gradient is as illustrated in Fig. 122.   As the gradient line more closely approaches the vertical, for bottom-poured castings, the conditions become more favorable for producing soundness by superimposed feedheads.   It is not suggested that the gradient lines shown on Fig. 122 accurately represent the metal temperature of a casting; they are merely indicative of the general trend of the bottom-pouring effect.

The transfer of heat from the metal to the mold is the prime reason for the notable difference in temperature existent between the upper and lower metal in the mold.   That portion of the lower level of the mold closely adjacent to the ingate is heated considerably and delays the solidification of the steel.   In some cases this adverse temperature gradient is corrected, or partially corrected, by running an auxiliary gate into the

riser, shown at C in Fig. 121, or by pouring directly down into the riser after the metal has risen into the bottom portion or neck of the riser.

A temperature gradient of the type postulated in Fig. 123 would probably be produced a short time after pouring was completed. The temperature gradient is not so steep as that exhibited by Fig. 122; nevertheless, it is still more than desired. The local favorable temperature gradient within the feedhead will ensure that the casting proper immediately below the riser will be sound; but the adverse nature of the temperature gradient in the lower portion of the mold, resultant upon bottom
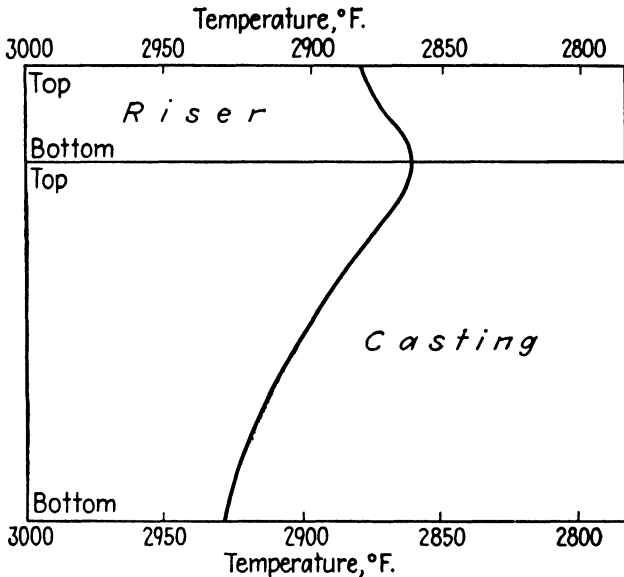
Fig. 123.—Temperature gradients in metal in bottom-gate systems when risers are top-filled with hot metal. (*Batty*.[16])

gating, points to the conclusion that the casting cannot be expected to be free from unsoundness in the lower levels.

In certain large castings, because of their design, it is possible to use bottom gating and the top filling of risers and to establish the proper thermal gradient toward the risers. Examples illustrating the employment of bottom gating and top filling of risers are anchor crowns and steel propellers.[38] Both of these castings, by design, are subject to directional solidification. The thin flukes on the anchor and the thin blades of the propeller solidify first, while the more massive sections, particularly the feedheads, remain fluid for a long time. The adverse temperature gradient produced by bottom pouring of these large castings would not only be short-lived but would soon reverse itself into a favorable one because of the low surface-area-to-volume ratio of the heavier base and

feedhead sections, and because of hot metal added to the risers.   In some
cases, when fast pouring is used from a number of bottom gates and the
sections are not of uniform thickness, it is possible to obtain the proper
temperature-gradient distribution by top filling of the risers.

There are three ways in which the disadvantages of bottom gating may
be overcome, at the same time preventing mold erosion by falling molten
steel.   These methods are (1) mold-reversal manipulation, (2) gating into
blind risers, and (3) step gating.

Mold-reversal Manipulation.—Mold-reversal manipulation allows
bottom gating of castings; by changing the position of the mold from 30
to 180 deg. in the vertical plane, the principle of top feeding can be used.
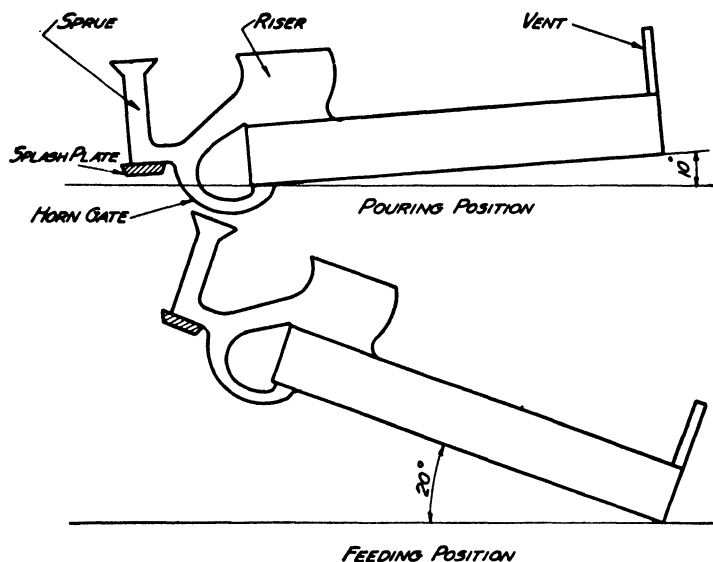
Fig. 124.—Partial-reversal manipulation.   (*Batty.*[15])

The gating system shown in Fig. 124 is devised to limit the flow of
metal through the bottom ingate and to cause metal to flow through the
upper sprue into the riser.   This provision helps to ensure the heating of
the riser cavity so that the proper temperature gradients of hot metal, hot
mold at the riser, and relatively cold metal at the point farthest from the
gate are established.   A pouring angle of 10 deg. has been found satis-
factory for proper bottom gating.   This angle also permits the metal to
rise as a unit instead of spreading over the entire mold cavity.   The cast-
ing is reversed through 30 deg. after pouring, to allow for gravity feeding.

The most favorable temperature gradient in both metal and mold in
bottom-pouring practice may be attained by a system of operations that
has been called the "total-reversal method."   An example of this method

is shown in Fig. 76. The sprue is either of tile or of cores. A number of ingates connect the annular riser to the annular distributor gate. The pouring operation, which has been predicated upon experience, applies metal as hot, and a pouring rate as slow, as practicable. These provisions produce in metal and mold the greatest practicable temperature gradients which, after the mold is reversed through an angle of 180 deg., allow the greatest possible feeding action.

Under certain cases the gating into blind risers, acting under atmospheric pressure, has been substituted for mold-reversal-manipulation methods (see Figs. 78 and 152). The substitution is beneficial, for it
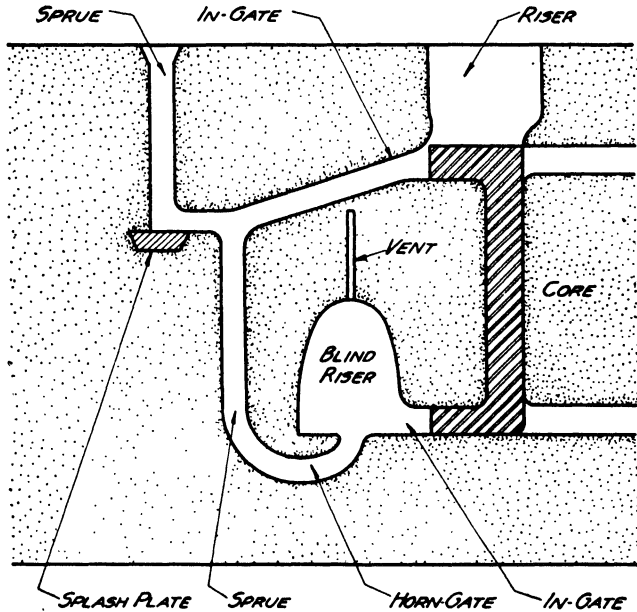
Fig. 125.—Gating into blind risers.

means that the castings can be poured on the flat and do not require tilting after pouring.

Gating into Blind Risers.—It was stated that by the use of blind risers, advantage may be taken of the bottom-gating system. This is merely bringing the riser down to the bottom of the casting to feed the hot spot produced by the adverse temperature gradients resulting from bottom gating. In order to make the blind riser function properly, the sprue should enter directly into it. A short ingate can then be run from the base of the riser to the casting, as shown in Fig. 125. A horn gate into the bottom of a blind riser has distinct advantages over either the straight or tangential ingate, since the fountain effect of the horn gate almost entirely prevents the accumulation of relatively cool metal within the blind riser.

This effect could also be accomplished by gating into the top of the blind riser, though, strictly speaking, this arrangement would no longer constitute a typical bottom gate. Other methods of making blind risers more effective are discussed under Risers.

Step Gating.—The step gate also takes advantage of bottom gating and at the same time allows hot metal to enter directly into the riser. The purpose of the step gate is to prevent the formation of hot spots at the bottom of the casting, which occurs when all the metal for the mold enters through the bottom gate, as noted in connection with Fig. 119.

Top Gating.—The advantage of top gating is that all the metal enters the casting directly through the risers, and hence the proper temperature gradients are formed, thus permitting efficient feeding. By using this
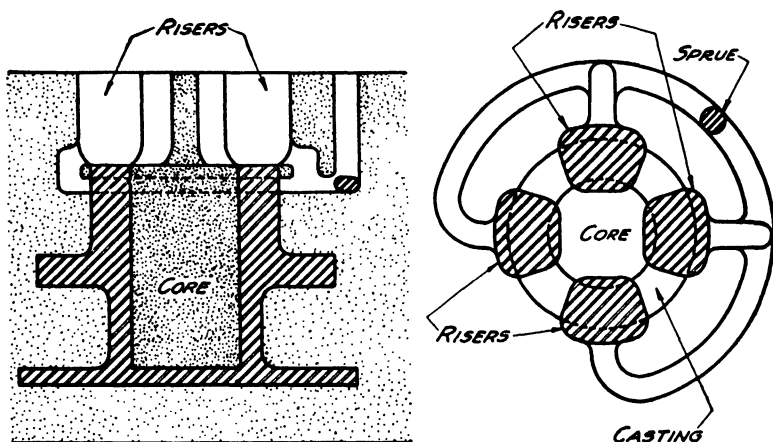
Fig. 126.—Top gating with use of a distributor or ring gate.   (*Johnson.*[34])

type of gating, directional solidification in the casting toward the risers is attained.

The disadvantage of top pouring is the erosion of the mold by dropping metal. Moreover, there is a serious possibility of the entrapment of air in the mold by top gating. Top gating can be effectively employed with a minimum of mold erosion only when there are a number of top ingates, so that their effective cross-sectional area is greater than that of the sprue (see Fig. 126). If this is done, the metal will enter the mold quietly and softly. A precaution that must be observed in top gating is to make certain that the dropping metal does not impinge upon cores, since this will add considerably to mold erosion. If possible, the gate should be constructed so that the metal runs down the mold wall instead of along the core face. A step-up gate into the risers is beneficial, for this arrangement prevents the risers nearest the sprue from receiving the greater portion of the metal.

Theory of Gating.—It is now proposed to examine in some detail the factors that govern the different types of running used with each of the three chief gating methods. They are parting-joint gating, bottom gating, and top gating. In all the calculations presented, it is assumed that steel is a perfect liquid and follows the laws of hydraulic flow. For the purpose of clarity, such losses as friction and change of section and direction in the various gates will be neglected. The effect of these items on the flow of steel will be considered later.

The results given by the calculations are, of course, purely theoretical. Nevertheless, they give useful information as to the relative effects of the various gating methods, and they present a clearer conception of the rules of application.
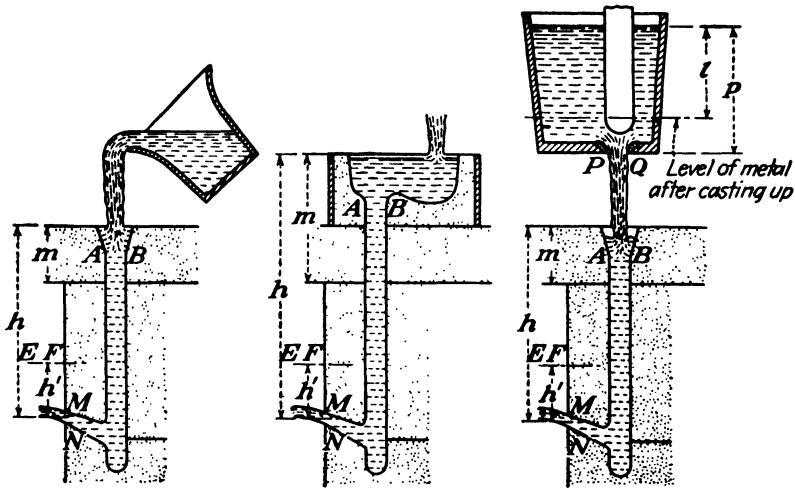
Fig. 127.—Molds cast under pressure conditions with the use of parting gates. (*Henon.*[22])

Casting under Pressure Conditions.—In Fig. 127, three types of pouring method are shown on similar castings that are gated on the parting. The diameter of the ingates as they enter the casting is smaller than the diameter of the sprues, and thus the molds are cast under pressure.

Since the ingate is smaller than the sprue, the metal accumulates at the bottom of the downgate and starts to run into the mold cavity. The amount of steel entering the sprue from the ladle is greater than that entering the mold from the ingate. As the sprue becomes full, it is necessary to slow down pouring from the ladle so that the metal input into the sprue equals the amount discharged from the ingate.

At the moment that the pouring basin is filled, the sprue and basin together can be regarded as a reservoir of metal maintained at a constant level in which the speed of arrival of the steel from the ladle is without influence. The height of the metal poured from the ladle is unimportant

in figuring the pressure head, since the metal falls into a pouring cup or basin and the distance h can be considered as the effective pouring head. Thus at the opening MN the flow is under pressure h.

The velocity with which the steel enters the mold cavity at the opening MN is equivalent to

$$V = \sqrt{2gh}$$

where V is the velocity of flow in feet per second, h is the height of the metal column or the pressure head, and g represents the gravitational acceleration.

Since
$$\sqrt{2g} = 8,$$
$$V = 8 \sqrt{h} \text{ ft. per sec.}$$

If S equals the cross-sectional area, in square feet, of the ingate at the point it enters the mold, then the quantity Q of steel that enters the mold will be

$$Q = V \cdot S = 8S \sqrt{h} \text{ cu. ft. per sec.}$$

The speed with which the metal enters the mold remains constant until the moment the mold fills to a level with the ingate MN. From
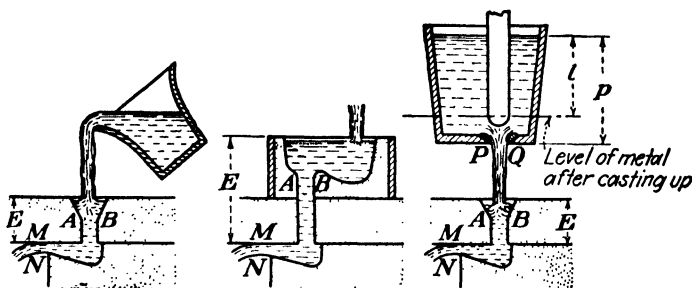
Fig. 128.—Molds cast under pressure conditions with the use of top gates. (*Henon.*[32])

this moment the metal flow constantly decreases in speed; thus, assuming the level of the metal in the mold cavity EF to be at a height h' from the gravity center of the opening MN, then the velocity of metal flow into the mold cavity becomes

$$V = 8 \sqrt{h - h'}$$

If the castings were bottom gated, the values given for gating at the mold parting likewise hold. Of course, from the start of pouring the opening MN is closed by the rising metal. The input therefore constantly decreases as the speed of entry of the steel into the mold cavity decreases.

If the ingate is located at the top of the mold cavity, then top-pouring conditions such as illustrated in Fig. 128 become active. The height of

the ingate MN is generally of a low order in relation to the height E. As it is negligible, it may be postulated that the velocity of the entering steel is

$$V = 8 \sqrt{E} \text{ ft. per sec.}$$

and the quantity of steel entering is

$$Q = 8S \sqrt{E} \text{ cu. ft. per sec.}$$

The velocity of the entering metal and the quantity of input are constant during the filling of the mold cavity, under the condition that the input from the ladle is also constant.
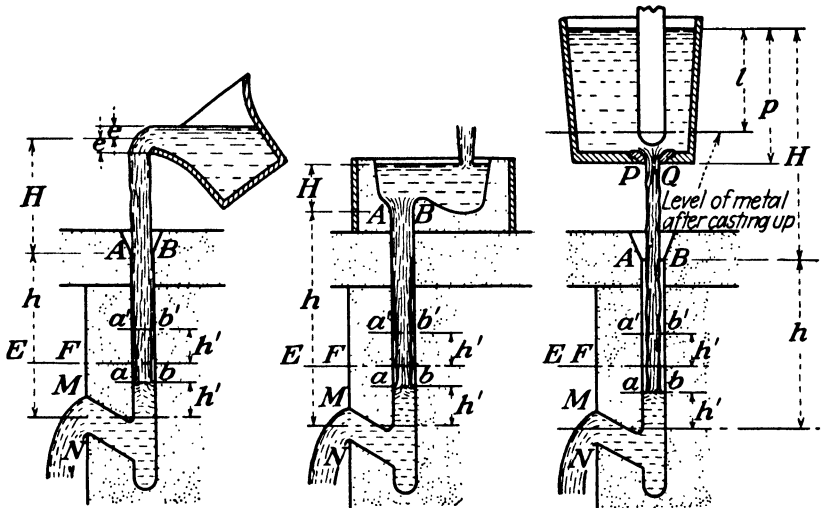
Fig. 129.—Casting under subnormal pressure conditions, gated at the parting joint. (*Henon.*[32])

Casting under Subnormal Pressure Conditions.—If the ingate is larger than the sprue, the metal will flow into the mold under subnormal pressure. This is illustrated in Fig. 129 for castings that are gated at the parting joint. This case is somewhat different from that shown in Fig. 127 because the pouring basin is not kept at a constant level and the ingate is larger than the sprue.

From the beginning of the casting the stream of metal exactly fills the orifice AB in the cases of the top-pour ladle and the basin-filled methods. The velocity of the steel and the quantity input at the orifice AB of the sprue are

$$V_{AB} = 8 \sqrt{H} \text{ ft. per sec.}$$
$$Q_{AB} = 8S_{AB} \sqrt{H} \text{ cu. ft. per sec.}$$

In the case of bottom pouring, the stream fills the nozzle PQ, and the velocity of the metal and the quantity that leaves the ladle are

$$V_{PQ} = 8 \sqrt{P}$$
$$Q_{PQ} = 8S_{PQ} \sqrt{P}$$

Now the metal drops down the sprue to the bottom of the well; it collects there and builds up and flows into the mold cavity by the ingate MN. Assuming that the speed of the stream of metal is completely interrupted on arrival at the bottom of the sprue, then it is necessary for the metal to build up to ab in the sprue before a full flow is obtained from the ingate MN.

The speed of flow of the metal into the casting and the quantity of metal will be equivalent to

$$V_{MN} = 8 \sqrt{h'}$$
$$Q_{MN} = 8S_{MN} \sqrt{h'}$$

where S_{MN} equals the cross-sectional area of the ingate MN.

Thus the speed, or velocity, of the metal entering the mold depends solely upon the height h'. Of course, the height h' depends upon the cross-sectional area of the ingate MN.

In the subnormal pressure casting a rate of entry of the metal into the mold cavity as low as is desired can be obtained if the shape of the casting allows for a large opening ingate. With a bottom-pouring ladle, the input being a variable, the distance h' and the velocity V_{MN} also vary.

These conditions persist until the moment at which the level of the metal in the mold cavity reaches the opening MN. From this moment onward the level of the metal in the sprue also rises. Assuming the metal at EF in the mold cavity, then it will reach $a'b'$ in the downgate at a height above that of EF, which will always be equal to h', for the flow in the section MN is governed by

$$Q_{MN} = 8S_{MN} \sqrt{h'}$$

Thus the speed V_{MN} remains constant during the pouring except when a bottom-pour ladle is used, at which time it decreases. The quantity of metal leaving the ladle in a definite time interval will vary according to the formula

$$Q = 8S \sqrt{P - l}$$

where P is the height of the metal in the ladle at the start of pouring of the casting, l is the height at the end of pouring the casting, and S is the cross-sectional area of the ladle nozzle.

Circumstances regarding the gating at the parting joint apply equally well to the bottom gating of castings, except that the opening MN is closed by metal from the beginning of casting.

In the top pouring of molds it is seldom that but one large ingate is used. Generally a number of small ingates are constructed so that the sum of their cross-sectional area is greater than that of the runner gate. In this case, conditions similar to those found in the use of the parting gate are encountered, with the exception that the metal does not rise over the gates until it enters the risers.

Nature of Metal Streams.—If the velocity of a metal flowing through a gate could be determined at two points 1 and 2 separated in the gate by
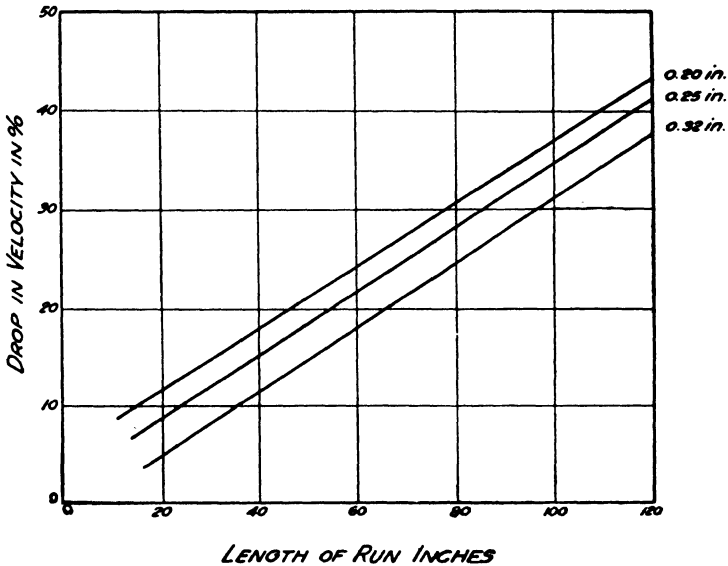
FIG. 130.—Drop in velocity of run with size of channel. (*Ruff.*[22])

a distance L, the drop in velocity S_v between the two points in question may be calculated for distance L to be $S_v = V_1 - V_2$. Generally this quantity is expressed in the form of a percentage.

An interesting point observed by Ruff[22] for steel is that the drop in velocity increases with the rate of flow. For example, in Fig. 130, it is observed that for any constant length of gate, a gate of decreasing diameter will show a higher drop in velocity. This is predicated upon the use of a constant-size sprue larger than the ingate. Hence, as the ingate is reduced in dimensions, the velocity or rate of flow of the metal in the ingate increases. The increase in the drop in velocity with the rate of flow is not what one would expect, since the longer the metal takes to flow over a given distance, the greater the amount of heat lost and the sooner, therefore, should crystallization ensue with accompanying increase in the

resistance to flow.   If, then, no such increase in resistance does in fact occur, the only explanation appears to be an increase in the internal friction of the metal at higher velocities of flow.

Figure 130 shows clearly that a drop in velocity of the flowing metal increases in proportion to the length of run and, conversely, that it decreases in proportion as the diameter of the flow becomes greater.   It should be mentioned, of course, that changes in the direction of a gate are responsible for variations in the velocity of the metal passing through the gate.
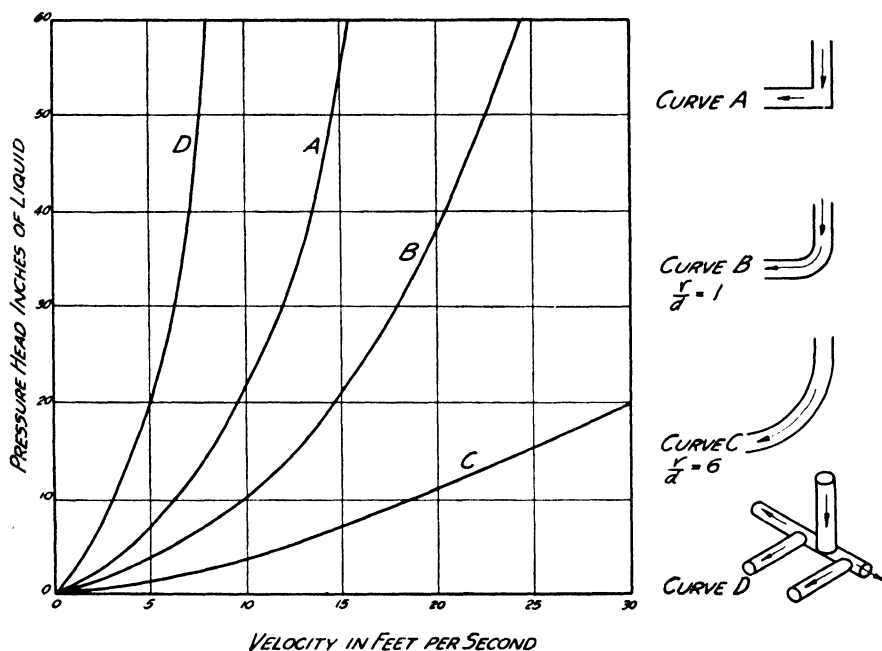
FIG. 131.—Changes in gate direction and their effect on velocity of run.   (Lips.[28])

Figure 131, as prepared by Lips,[28] shows the variation of the velocity of metal flowing under pressure through various designs of ingate. Curves A, B, and C show the loss of velocity that occurs when the molten metal undergoes a change in direction of 90 deg.   Curve D shows what the losses may be when there are two sudden changes in direction and a sudden increase in the cross section of the ingates.

Ruff[22] observed that the drop in velocity existing between two points in a channel or ingate decreased as the temperature of the steel decreased. This condition also is contrary to the generally accepted opinion, for it is generally thought that as the temperature of the molten steel is increased, the viscosity of the steel is lessened and hence the retardation in velocity over a definite distance would be at a minimum.   Just what is the cause

for this effect is not definitely known, but it may be the result of internal friction of the metal, increased quantities of gas, or changes in the turbulence of the metal, with the last named probably the most important.

It has been found that the nature of the flow of water through pipes differs, depending upon the velocity of the water and the diameter of the pipe. Studies have been made that show that under certain conditions dye added to the center of a flowing stream will be carried along the length of the stream without being mixed with the remainder of the stream. When such conditions occur, it is said that the flow is streamlined or laminar. In this case a very steady and smooth-exit flow occurs. If, however, the velocity of the water were increased, a point would be reached when the dye would no longer occupy the center portion of the flowing stream but would become dispersed, and the entire stream would become dyed. An unsteady flow of this type is known as a "turbulent" flow.

The character of the flow of water as to whether it is laminar or turbulent is expressed by its Reynolds number. The Reynolds number is equivalent to the following:

Reynolds number $= R_n =$

$$\frac{\text{mean velocity of flow in pipe} \times \text{diameter of pipe}}{\text{kinematic viscosity of the liquid}}$$

It has also been found that if the Reynolds number describing the flow of water in a pipe has a value less than 2,000, the flow will in general be laminar, whereas, if the Reynolds number is greater than 2,000, the flow will almost certainly be turbulent. The Reynolds number at which the flow changes from smooth or laminar flow to one of turbulence is known as the "critical velocity" of the stream. Irregularities in the pipe or increases in the diameter of the pipe such as tapering enlargements always lower the critical velocity.

The flow of metal is apparently similar to that of water, for laminar flow and turbulent flow are observed. The critical velocity, related to the gate diameter for cast iron and aluminum, has been reported upon by Lips.[28] For a gate ½ in. in diameter, the critical velocity for cast iron would be approximately 0.16 ft. per sec. (approximately 2 in. per sec.). This is an extremely slow-flowing metal. The critical velocity decreases as the diameter increases.

A study made by Ruff[22] showed that the flow of steel was always under turbulent conditions. Ruff plots the Reynolds number in relation to the resistence index, as is shown in Fig. 132. The resistance index is equivalent to

$$\lambda = \frac{2d}{V^2} a$$

Where d is the diameter of the pipe or gate, V is the discharge velocity of the metal and a is the retardation of the metal flow per unit length.

It will be noticed from Fig. 132 that the experimental points are all located in the zone of turbulent flow and, since these data are based on the flow of steel through 6- and 8-mm. channels, it can easily be seen that for gates larger than this size in diameter, the Reynolds number will be definitely higher and the flow will likewise be turbulent in nature. Estimates as made by Ruff for steel show the critical velocity for smooth flow to have a Reynolds number of approximately 3,500.

The turbulent motion of molten steel is shown in Fig. 133 for a gate completely filled with steel. This figure has been prepared by Ruff[22] to
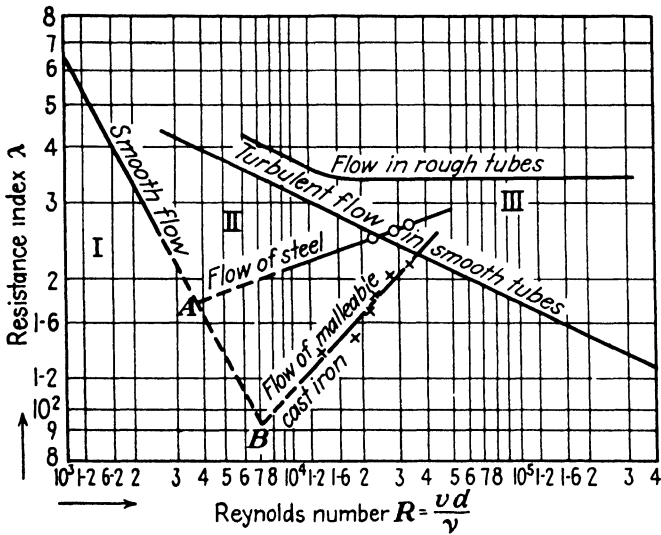
Fig. 132.—Resistance index in relation to the Reynolds number for steel. (*Ruff.*[22])

represent the way in which individual particles of liquid metal may be imagined to move in a gate. The particles arrive at the core of the flow with a maximum velocity; owing to the turbulence they are then thrown against the wall of the mold from which they finally again return to the center. Hence, on the way from the core to the wall their velocity, as measured in the direction of flow, must drop until it finally becomes nil. Between the core and the wall, two other velocities are to be recognized in addition to the core velocity V maximum. The first of these is the mean velocity of flow V_m, which, when applied to the whole cross section, serves to measure the total velocity of flow of the material. The second is the velocity of flow V_w in the neighborhood of the wall, which governs the length of time that the individual particles of liquid remain close to

the wall and therefore the amount of heat they lose before returning to the interior of the flow.

In accordance with the conditions of turbulent flow, it seems that certain stationary zones of steel act as protection to the walls of the mold. These zones do not participate to any appreciable extent in the flow of the steel, and therefore it is of little matter whether these stationary zones lining the walls remain liquid or consist of solid metal. The nominal diameter of the channel will, however, be reduced by the fraction that ceases in this way to be available for the movement of the metal.

Vmax.—VELOCITY OF FLOW AT CORE.

Vm——DISCHARGE VELOCITY.

Vw——VELOCITY OF FLOW NEAR MOLD WALL.

de ——DIAMETER OF FLOW CHANNEL.

di ——DIAMETER OF INNER FLOW.

δ ——THICKNESS OF STATIONARY ZONE NEAR MOLD WALL.
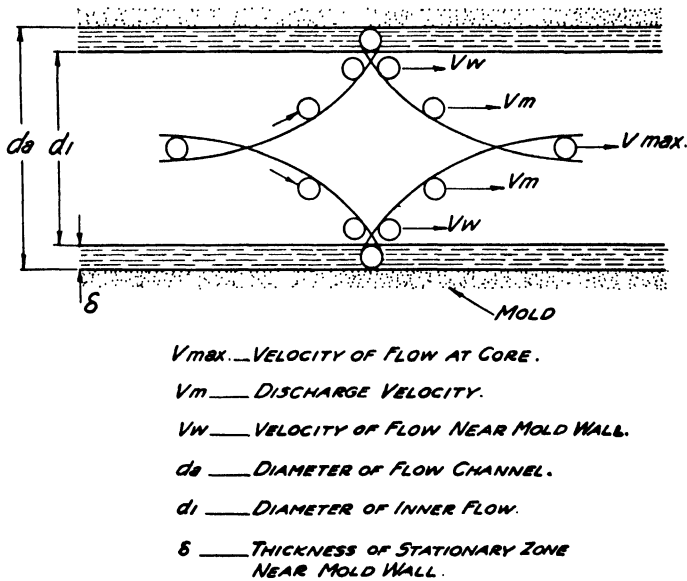
FIG. 133.—Mechanics of turbulent flow of molten steel. (*Ruff.*[22])

Size of Gates.—There appears to be but little information on the size of gates that should be employed on castings; experience is apparently the only criterion for acceptable practice. The fact that the casting must be successfully run has probably led to the use of gates that are more than adequate. This provides increased turbulence of the entering metal, which adds to mold erosion. This is especially true for the type of gating that permits casting under pressure.

A quietly flowing gate system for cast iron (as shown in Fig. 134) should have, according to Measkowski,[12] a ratio of sprue to cross gate to ingate of 4 : 3 : 2 or of 8 : 3 : 2, depending upon whether the metal flow is unidirectional or divided in the cross gate. A gate of this type, however, is seldom used for gating steel castings.

If thin castings are to be run,[31] they should preferably be poured on an inclined plane such as shown in Fig. 135.

It is necessary that any "resting" of the metal be avoided since there is always a risk of premature solidification.   Diagram A of Fig. 135 shows

$Sw : Sz : S = 4 : 3 : 2$ $Sw : Sz : S = 8 : 3 : 2$
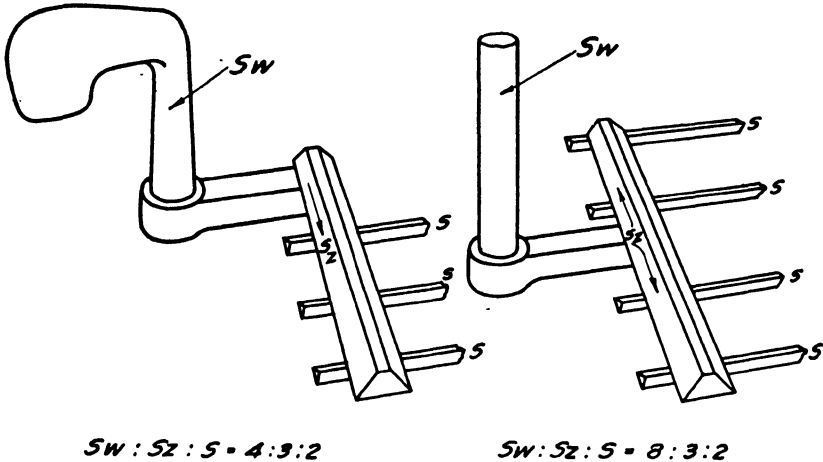
FIG. 134.—A quietly flowing gate system for cast iron.   (*Measkowski.*[12])

A   B
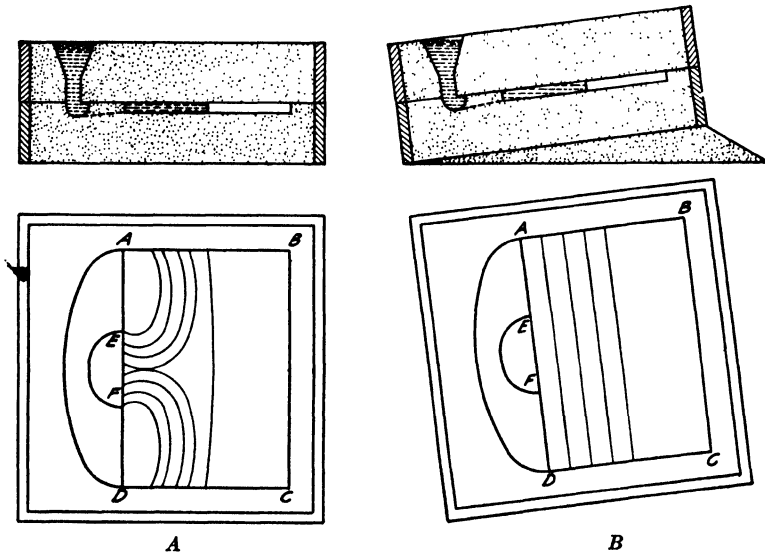
FIG. 135.—Flow characteristics on flat and inclined pouring of thin castings.

a plate *ABCD* cast horizontally.   The metal enters by the gates *AE* and *DF*.   The metal successfully fills the zones bounded by the curves 1, 2, 3, and 4.   After curve 1, the metal has a tendency to flow toward the center more than straight in front.   The low rate of progress of the

front edge of the metal may bring about premature solidification, and the casting shows either cold shuts or a short run. It is customary to overcome this drawback by slightly tilting the mold, as shown in sketch *B*. The metal immediately leaving the gates spreads itself along the length of the mold cavity and continues to fill the mold without resting at any point.

Very large castings are rarely top-gated, since the mold cannot withstand the cutting action or the impact of a vertically falling column of steel. In some cases, however, the metal is allowed to fall on metal splash plates or on refractory brick, which can better withstand metal erosion.

Parting gates are considerably less damaging in their effect upon the mold. In fact, the tendency to mold erosion is almost entirely eliminated when several of these gates simultaneously supply the metal to the mold. Multiple gating slows down the speed of the metal, causing it to run down the banks of the mold in slowly descending streams. Parting gates of the multiple type are shown in Fig. 80.

Bottom gating, despite the adverse temperature gradients it creates, enjoys the greater popularity among steel foundrymen. Its widespread popularity depends upon its ability to introduce metal into the mold with a minimum of mold disturbance. The step gating of large castings such as shown in Fig. 84 takes advantage of bottom gating as well as approaching the advantages of parting gates and top gates.

Pouring Speed.—A formula for determining the duration of pouring certain weights of castings has been proposed by Dietert.[8] Although the formula was applied to cast iron, it is applicable to any metal. According to Dietert,

$$\text{Pouring time in seconds} = S \sqrt{W}$$

where W is the weight of casting in pounds and S is a constant depending upon the type of metal and thickness of section.
For cast iron,

$$S = 1.1 \text{ for } \tfrac{7}{64}\text{- to } \tfrac{9}{64}\text{-in. section}$$
$$1.25 \text{ for } \tfrac{10}{64}\text{- to } \tfrac{20}{64}\text{-in. section}$$
$$1.5 \text{ for } \tfrac{21}{64}\text{- to } \tfrac{39}{64}\text{-in. section}$$

In order that the correct pouring time may be employed, formulas are given for the effective sprue height and the area of the choke.

According to Ronceray,[9] formulas for pouring time lead to grave errors when an attempt is made to generalize them, for not only the length of time of pouring must be changed with the nature of the castings, but due consideration must be given to the method of gating and feeding. An average speed of pouring for most large castings is equivalent to approximately 1.0-in. rise of metal in the mold per sec.

The rate of rise of steel in the mold is a subject that has been given too little consideration in the production of steel castings, for it may affect (1) the temperature gradients in the metal and mold and (2) surface characteristics of the casting owing to the heat action of the rising metal on the mold.

It has been observed that pilot castings can be produced without defects, whereas some of the subsequent castings molded and gated similarly to the pilot castings will show shrinkage cavities and/or hot tears. The reason for this is that no consideration has been given to controlling effectively the rate of rise of steel in the mold and the temperature of the steel. In order to control the rate of rise of metal in the mold, the size of the gates and method of entry must be more closely controlled than is the case in industry today.

Preparation of Gates.—Gates for steel castings should be part of the pattern equipment, since their location and size are of such importance to the casting. Besides, better mold surfaces can be obtained by ramming up the gates than by cutting them after the mold has been prepared. Rammed-up gates are true to size and shape and thus uniform in subsequent molds. Cut gates may or may not be uniform, thus adding to the number of variables in the casting practice.

Gates in some molds (nearly all dry-sand molds and some green-sand molds) are given special surface washes to make them more resistant to the metal erosion. Terra-cotta sleeves or hard-baked cores are frequently used for gates. These materials make excellent gates since they keep mold erosion to a minimum. In general, steel foundry practice calls for their use on most large- and medium-sized castings.

Some foundries prefer to use a core at the junction of the gate and the mold. The purpose of this core is to reduce the cross section of the gate so that the gate may be broken off by a sledge rather than cut off by flame cutting. A core used in this manner is known as the "Washburn core," and such a gate is sometimes referred to as the "Washburn gate."

Pouring Cups.—There are probably as many designs of pouring cups as there are foundries using them. A few general principles must be considered. The cup diameter should be about 2.5 to 3 times that of the sprue diameter, and the height should be equal to the diameter. The inside walls of the cup should be straight or at a steep angle. The latter condition is the one usually selected, for it is an aid in molding the cup. Cups for small-diameter sprues generally require shoulders, as shown in Fig. 136, so that the cup will have sufficient depth.

The main precaution is to have the diameter of the cup large enough. With too small a cup diameter, it may be impossible to pour sufficient metal into the cup to fill completely the sprue orifice in time to prevent dirt or slag from the ladle from flowing down the sprue. It is difficult

to fill a shallow cup without splashing and is more difficult to keep it filled during pouring. Two other designs of pouring cups are shown in Fig. 137.
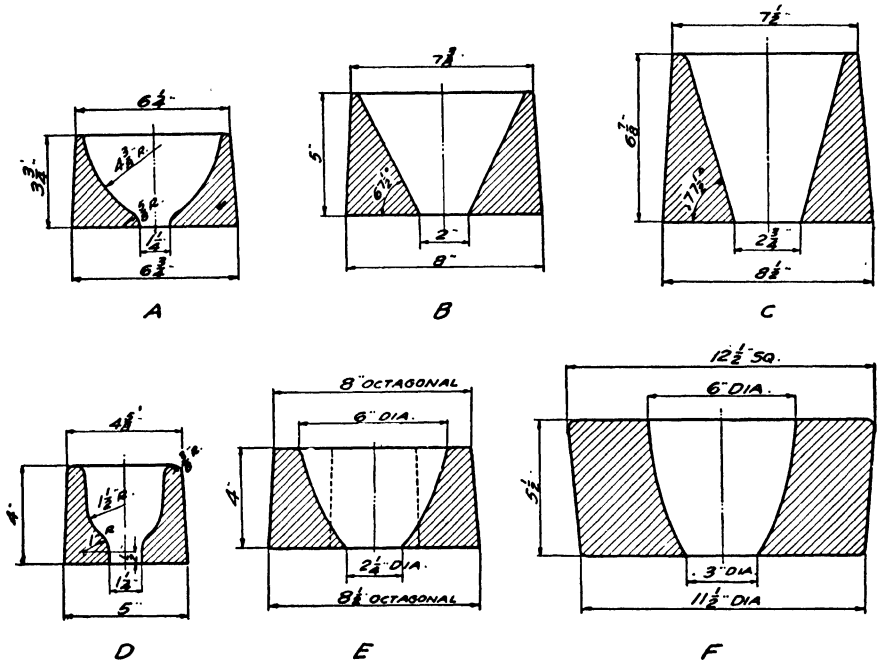
Fig. 136.—Designs of small pouring cups.
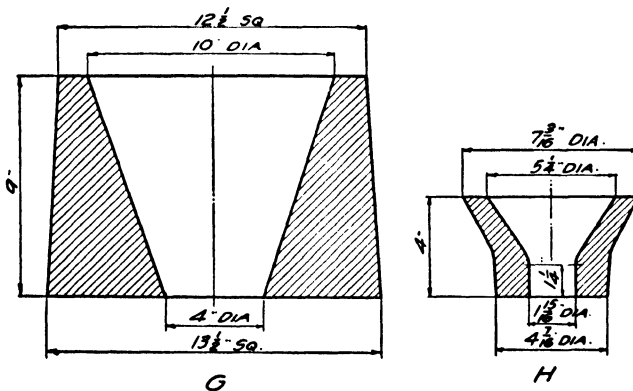
Fig. 137.—Designs of large pouring cups.

A pouring cup for top pouring is shown in Fig. 138. The distance between the shower gates will depend upon the distance between risers, since the pouring is usually done down two risers.

In nearly all cases pouring cups or basins are constructed from heap sand, to which have been added bonding materials for producing a hard-surfaced cup after oven baking.   In most cases, the inside surface of the basin is washed or coated to make it more resistant to metal erosion.

SHOWER GATES

POURING CUP
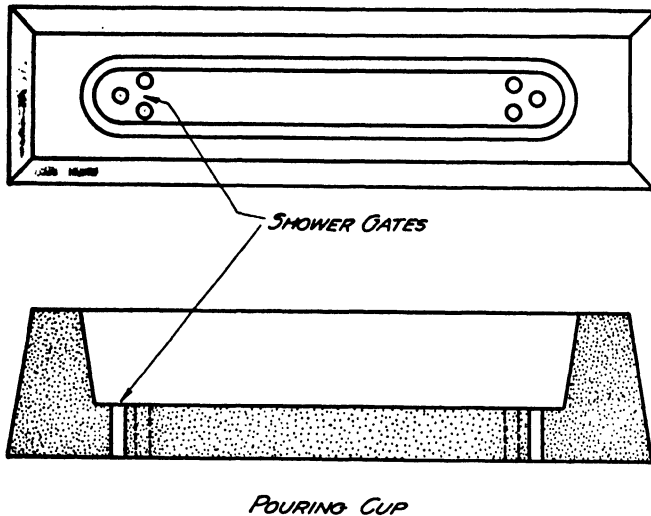
Fig. 138.—Pouring runner cup.

RISERS

The terms "riser" and "feedhead" refer to a reservoir of molten steel intended to satisfy the demand for feed metal in the casting proper, because of volume contraction of steel in the molten state and upon its solidification.   In discussing risers, it is not the purpose to explain the principle of casting solidification, but rather to discuss the shape and size of risers and the manufacturing details.

There are five primary requirements of risers:

1. To be effective, a riser must be the last portion of the casting to solidify.

2. A riser must cover completely the casting section that requires feeding.

3. The riser volume must be sufficient to compensate for the metal contraction within the casting.

4. The fluidity of the metal in the riser must be maintained so that the metal can flow from it and penetrate to the last contraction cavity.

5. The riser should be effective in establishing a pronounced temperature gradient within the casting, so that the casting will solidify directionally toward the riser.

It is to be observed, therefore, that the shape, size, and position of the riser must be effectively controlled.

Riser Types.—There are two types of riser: (1) open, and (2) blind. The term "open riser" refers to a riser the topmost surface of which is open to the atmosphere. A "blind riser" is entirely surrounded by the mold except possibly for a vent that may extend from the riser to the top surface of the mold.

Open risers are usually placed on the top of castings and but seldom extend downward in the flask much below the parting line. The force of gravity, as well as atmospheric pressure on the molten metal, aids in moving the feed metal downward into the casting proper as the demand for such feed metal arises.

Blind risers are usually placed in the drag side of the mold, at positions where ingates would enter the casting under normal bottom-gating methods. The force of gravity is active in supplying feed metal; and if certain precautions are taken for the introduction of atmospheric pressure to the interior of the riser, this force is also available in supplying feed metal.

Riser Shape.—The shape of the riser might well be predicated upon what is known of the effect of shape on the solidification of steel ingots. The method of casting ingots, small end up, led to the formation of a long pipe (Fig. 139) that became much shorter and wider if the ingot was cast big end up. An observation of these facts led to another improvement, *i.e.*, the placing of a feeder head or hot top in the shape of a truncated ingot mold of the small-end-up type on top of the big-end-up mold. Thus an ingot-mold system was designed with a hot top that would feed as deeply as possible above a mold that solidified with a short, wide pipe. The result of this combination is that an ingot free from contraction activities is produced. Not only is this true when ingots are cast in metal mold with refractory hot tops, but sand-cast ingots show somewhat similar characteristics.

Conditions that are effective in the ingot-mold setup can be reproduced on smaller scales for steel castings. Of course, the section directly under the riser is seldom shaped like an ingot, but nevertheless the principle involved can be used on many occasions.

It has been observed[18] in numerous cases, however, that some foundrymen prefer the type of riser shown in the upper sketch of Fig. 140. This riser was so designed that it could be knocked off with a striking hammer. This type of riser incorporates the big-end-up principle; however, one of the primary requirements was violated in that the riser does not fully cover the area to be fed. Because removing the riser was more important than feeding the casting, contraction cavities or a secondary pipe developed within the casting. If the design of the riser were changed

to that shown in the lower sketch of Fig. 140, a better feeding system would be obtained and the riser would be better able to fulfill the primary requirements of a riser. It may sometimes be advisable to increase the width of the mold cavity at the junction of the riser AA, so that the straight sections may be more properly fed.
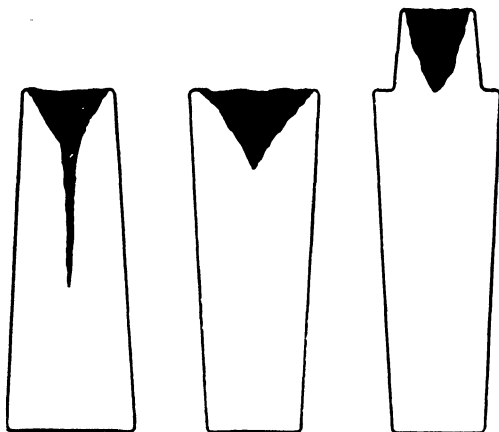
Fig. 139.—Pipe formation in ingots.
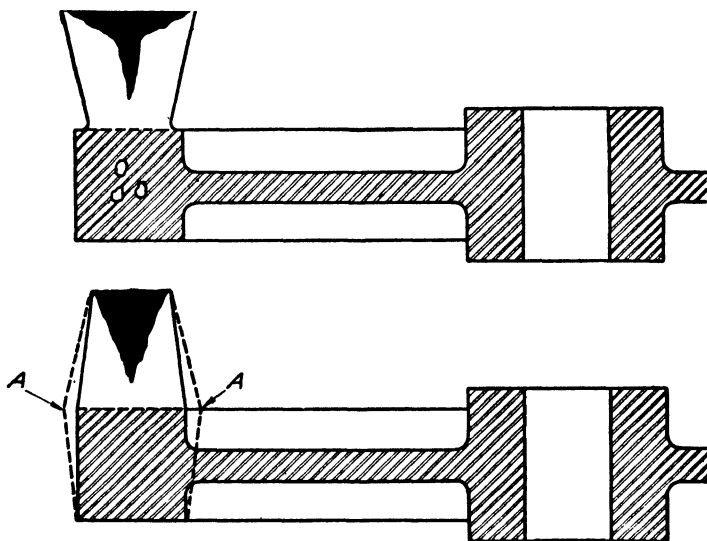
Fig. 140.—Ingot heading design for steel casting risers. (*Bennett.*[18])

The small-end-up type of riser presents but a small area at the top, exposed to air. This design has the advantage of bringing a large proportion of the metal in the risers close to the casting, where it has a better chance to keep the necks hot and retard the rate of skin formation. The maximum section of metal is deeply covered with sand, so that the metal

cools at the slowest possible rate; and the area exposed to the air is so small that the extent to which the metal is air-cooled and frozen at the top is greatly reduced. Likewise, the riser is best shaped to give the deep feeding desired.

A riser used in the steel foundry, and somewhat similar to that of Fig. 140, is shown in Fig. 141, sketch 1. This riser is often larger in cross section than the member on which it is placed, and it is usually connected to the casting by a neck thinner than either the riser or the casting section to be fed. The excuse given for the use of this type of riser is that it is
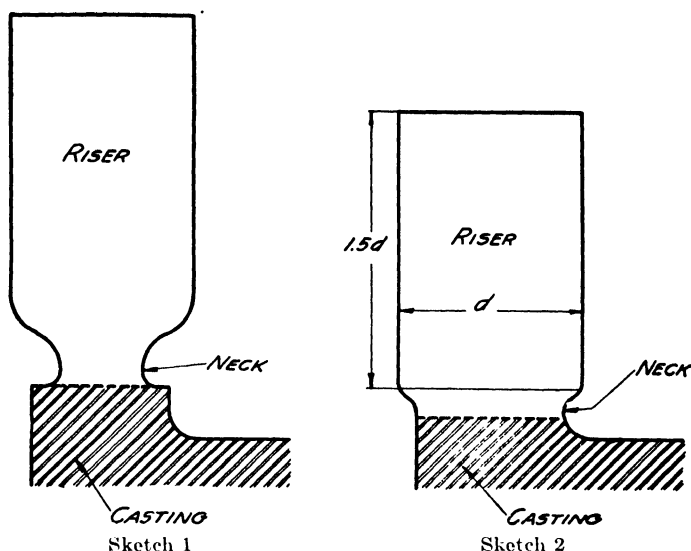
Sketch 1   Sketch 2
Fig. 141.—Incorrect and correct methods of constructing risers.
Sketch 1—Poor design
Sketch 2—Proper design

easily and cheaply broken off with the sledge, and the riser is high enough to permit the pressure of the metal to feed the casting effectively.

In the first place, it should be mentioned that risers can be cheaply and quickly cut off from the casting by the use of oxyacetylene torches. In the second place, gravitation of molten metal will take place in a riser only so long as liquid steel is present and a downward path is open. It has been suggested repeatedly that the height of a parallel-section riser is essential in order that benefit may be derived from liquid-metal pressure. If the riser were many times the height of the riser shown, no more efficient feeding of the casting would be obtained. Any amount of metal pressure cannot prevent the solidification of the neck portion as constructed in Fig. 141, Sketch 1, before the casting solidifies. A considerable amount of metal is wasted in risers of this shape.

A riser should be provided with a neck that is as wide as the section to be fed.   It should also be quite short as shown in Fig. 141, Sketch 2.   The riser should be only enough wider than the neck to make certain that it stays fluid at the center as long as there is any possibility of its feeding metal through the neck to the casting below.   The small quantity of metal in the neck is kept molten by the mass of hot metal in immediate contact with it in the riser.   An opinion is widely held, since it is based upon some experimental study, that the most effective height of risers is 1.5 times their diameter to produce maximum feeding for the minimum metal expenditure.

If a riser is shown by actual trial to be inadequate for feeding the casting to which it is attached, care should be taken in increasing the riser volume.   Assuming that the riser was well proportioned, any appreciable
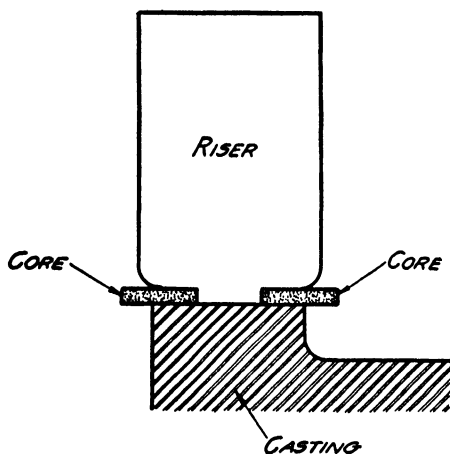
Fig. 142.—Cores at the base of a riser (necked-down riser).

increase in height should be accompanied by a proportionate increase in diameter, so that the thermal efficiency of the riser will be maintained or increased by the change.   If the height of the riser is raised without increasing the width, it is possible that a long secondary shrinkage cavity will develop, indicating poor riser design.   The importance of riser design is often overlooked when preparing castings for production.   Risers relatively long with respect to their diameter are inefficient in their use of feed metal and quite often fail to feed the casting properly.

There are times when it is possible to make use of a necked-down riser.   However, considerable study must be given to obtaining the proper proportions in such instances.   The success of the venture lies with the use of a very short neck, such as shown in Fig. 142.   In all cases a core plate is used to form the neck.   For risers of 6-in. diameter the thickness of the core is between ½ and 1 in.   The core heats to the metal

temperature; and, since sand is a relatively poor conductor of heat, the core does not cool the neck but prevents the neck from solidifying much ahead of the main body of the riser.

Some recent work on ratios between neck diameter, core thickness, and riser diameter shows that 4- and 6-in.-diameter risers will feed similar-sized castings through a 1-in.-diameter neck with a 1-in. core thickness. The ratio is apparently not a straight-line function but more nearly hyperbolic. The advantage in this type of riser is that removal from the casting is much easier. This point is worth considering in the torch cutting of risers 4 to 12 in. in diameter. The core used in the necked-
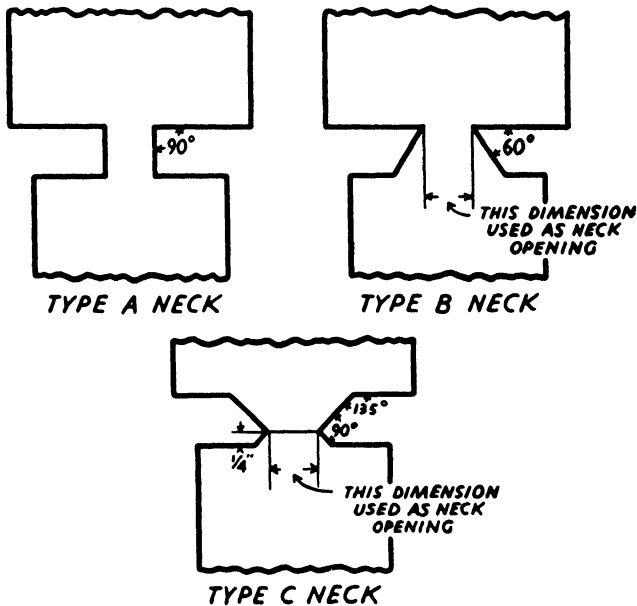
FIG. 143.—Neck designs for necked-down risers. (*Cain.*[48])

down riser is often referred to as the "Washburn core," because of its similarity to that type of core used in connection with gates. A study has been made[48] showing the size and shape of neck for castings 2, 4, and 6 in. in diameter. Figure 143 illustrates the design of necks used in the study, and Fig. 144 illustrates the minimum relation that can be used. Relationships to the left of any line on the graph will result in a solid, well-fed casting. Values that would plot to the right of any line will produce shrinkage in the casting. It will be noted that there is a definite relationship between the thickness of the necking core and the cross-sectional area of the neck opening. The design of the neck is also an important variable in determining this relationship. It is interesting to note that a very small neck will permit the efficient feeding of a casting.

Some consideration should be given to the gating of castings into risers, for by gating into the riser it is possible to create an optimum temperature gradient from the casting to the riser.   A tangential entry of the gate into the riser allows the incoming metal to preheat the walls of the riser cavity while the mold cavity is being filled and also cushions its entry into the mold.   If the ingate is constructed so that it is the full height of the riser, then the incoming metal will be at the top as the riser is filled.
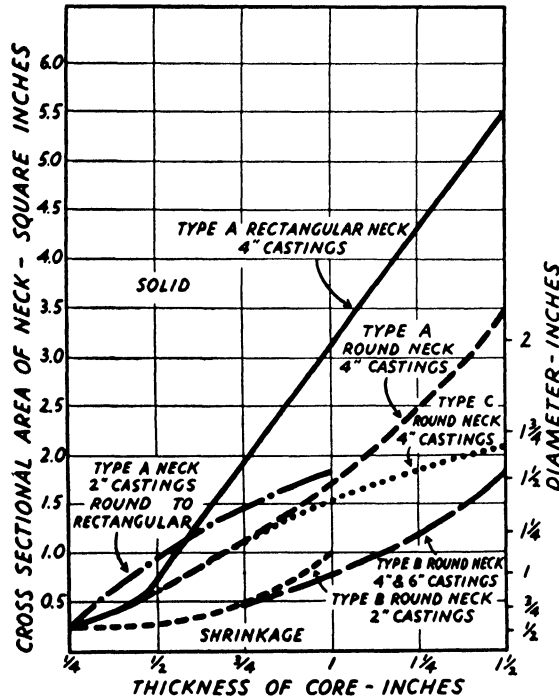
FIG. 144.—Relation between the cross-sectional area of the neck and the thickness of the core to produce castings without shrinkage.   (*Cain.*[48])

In this way the top of the riser metal is kept hot and in motion so that it will not solidify.

Risers may be elliptical, annular, square, or irregularly shaped, according to conditions.   If it is necessary to make them square or rectangular where they join the casting, the shape may be changed so that they become cylindrical above the junction.

The best efficiency of a riser is obtained when the ratio of its volume to the surface is at a maximum.   Thus the ideal form of a riser is a sphere, but there are practical difficulties in economically applying such a form of feeder.   For a volume the same as that of a sphere, the next most efficient form of riser is the cylinder; following this is the square and then

the flat, rectangular types of riser. To illustrate by way of a concrete example, Table XLIII shows the relative feeding efficiencies of differently shaped risers of identical volumes.

TABLE XLIII.—FEEDING EFFICIENCIES OF DIFFERENTLY SHAPED RISERS OF IDENTICAL VOLUMES

Form and size of riser	Vol., cu. in.	Weight, lb.	Area, sq. in.	Amount solidified in 1 min., lb. per cu. in.	Time to solidify completely, min.
Sphere: 6-in. diam................	113	32.0	100	42.7	7.2
Cylinder: 4¼ by 8 in.............	113	32.0	120	51.2	4.7
Square: 3⅝ by 3⅝ by 8⅝........	113	32.0	135	57.5	3.6
Plate: 2¼ by 6¼ by 8 in.........	113	32.0	160	68.4	2.7
Plate: 1²⁵⁄₆₄ by 10⁵⁄₃₂ by 8 in......	113	32.0	220	93.8	1.5

Molten steel present in the corners of square and rectangular risers solidifies very early in the period of solidification owing to the maximum amount of surface cooling offered by the mold.

Weight of Risers.—Risers greatly influence the cost of the finished casting, since for small and intricate castings the weight of risers may vary from 10 to 50 per cent of the casting, while for larger castings the weight of the risers may be equal to the weight of the casting. Moreover, the risers must be removed and the casting shaped to the proper contour before it can be shipped to the buyer. Thus the weight of risers, the positioning of them, and the area of contact between the riser and the casting are important. Also, since the size and number of risers determine the amount of metal in excess of that comprising the casting proper, it is clearly desirable to use as few and as small risers as possible, commensurate with soundness in the casting.

It is known that the liquid contraction of steel is approximately 1.6 per cent per 100°C., and that the contraction on solidification is about 3 per cent, giving a total of about 5 per cent contraction from the time the molten metal is poured into the mold until it completes solidification. In view of these values, why is it that the weight of risers must be so excessive? In answering this, it must be remembered that steel solidifies from the mold-metal interface toward the center of the section, and that the last solidifying position is the place where the volume contraction appears in the form of cavities. To prevent this cavity formation, the riser must furnish a supply of liquid metal. However, the riser has been solidifying at practically the same rate as the casting, but because of its shape and size it contains an amount of liquid metal in its center sufficient

to prevent the formation of the contraction cavities.   In other words, a large amount of metal must be present in the risers to safeguard the availability of the small amount needed to supply metal lost through contraction.

There are times when the use of unusually large risers is necessary, *viz.*, when their form is such that they have a low thermal efficiency, when they are intended as reservoirs of temperature to direct solidification, and when the cope selected for the job is too high.

Favorable temperature gradients are produced by risers only when their ratio of surface area to volume is considerably less than the ratio of surface area to volume of the casting that they surmount.   This is predicated upon the fact that the velocity of solidification is a constant, *i.e.*, at any definite time after pouring, the weight or volume of steel solidified per square inch of mold surface is the same.

Adverse temperature gradients may be set up in a mold by bottom pouring.   Any attempt made to correct this adverse gradient would necessitate employing extra risers, especially if the risers were filled with metal that had traversed the entire mold cavity.   If, however, the risers were top-poured with hot metal, the adverse temperature gradient will not be so great.   These conditions are shown in Figs. 122 and 123.   It is the policy of steel foundrymen to add hot metal to risers whenever the casting is of sufficient size to warrant this practice.

Some operators appear to have a fondness for filling risers of castings after considerable shrinkage has taken place, but it is believed that such procedure is almost worthless as applied to the range of medium-weight castings between 500 and 2,000 lb. each, particularly in the case when the bottom-gating procedure is followed.   The metal in these risers is so cool that the additional metal merely fills the shrinkage cavity and has but little effect on increasing the temperature at the neck of the riser to make it more effective.   In small castings most sections have completed their solidification before a time interval of 3 or 4 min. has elapsed.   The repeated feeding of risers of very large castings is entirely another matter, and such procedure is quite effective.

In the making of small- and medium-weight castings, the riser cavities should be constructed to such a size that they will contain all the metal essential to complete feeding when filled to within 1 in. of the top at the time of pouring.   The surface should then be immediately covered with an insulating material.

It is not advisable to fill risers even with the surface level of the top of the mold, since this prevents covering the top of the cast metal with a sufficient thickness of material to serve as an insulator.   Metal exposed to the atmosphere loses its heat more quickly to the atmosphere than it does to the mold.   When this occurs at the surface of a riser, an appre-

ciable proportion of what should be feed metal is immobilized by surface solidification. Also, a partial vacuum is formed under this crust, and the riser loses some of its effectiveness to feed downward.

The usual practice of pouring metal into risers is not entirely effective and can be improved. If a casting has three risers, such as shown in Fig.
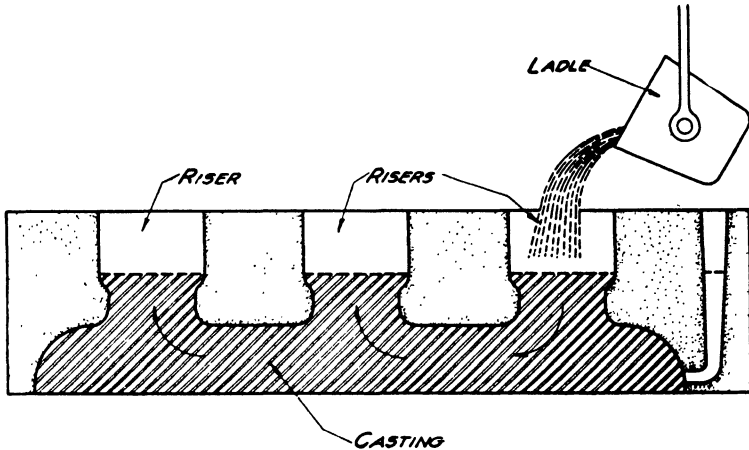
Fig. 145.—Usual method of top-filling risers.
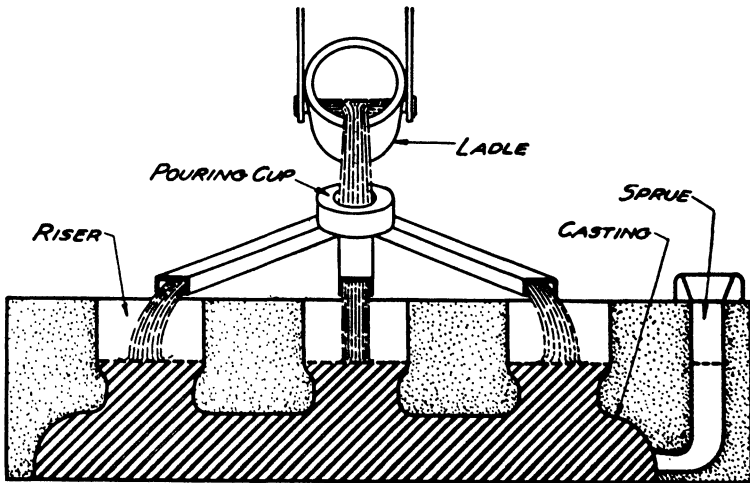
Fig. 146.—Best acceptable method of pouring into risers.

145, the usual method of pouring down the risers is to pour one riser until filled. In so doing, the other two risers are filled with relatively cool metal from the mold cavity and not hot metal from the ladle. All three risers may be filled with hot metal if, instead of pouring into one riser, the metal is poured into a gate that crosses the top of the mold to the three risers, such as shown in Fig. 146. Metal poured into the gate will fill all

the risers simultaneously with hot metal, and the correct temperature
gradients are established in each riser to allow for more effective feeding.

The proportion of feed metal that has to be provided for each casting
is a variable depending upon the design of the castings. In fact, the
proportion of feed metal that must be supplied to different risers on a
single casting is also a variable depending upon their position. For
example, Batty[15] found that feed metal applied to the rim of a gear blank
is adequate when it is equal in volume to between 16 and 20 per cent of
the volume of the rim plus spokes or plate. The volume of the riser
imposed on the hub varies with the interrelated dimensions of the hub.
The proportion applied to the hub is necessarily greater than that applied
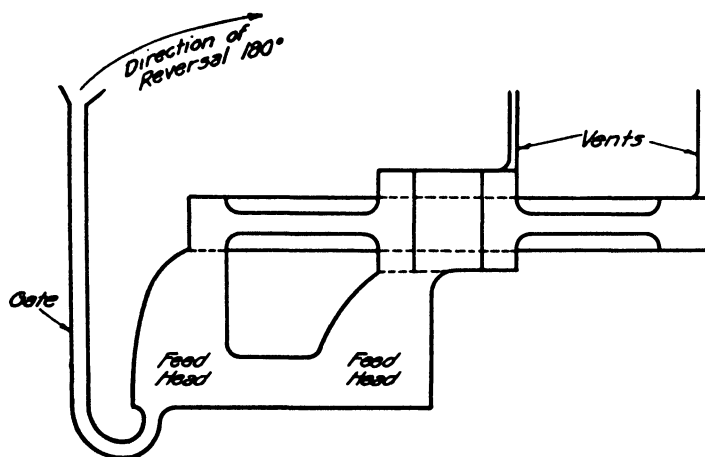
FIG. 147.—Gear-blank production by the total-reversal method. (*Batty.*[15])

to the rim, because the metal that fills the hub riser has been cooled in
transit through the mold and is superimposed on metal a little hotter than
itself. In the hub and hub riser, there is an adverse gradient, and it is
therefore necessary to have a hub riser of a size that appears dispropor-
tionate as compared with the riser imposed on the rim. As a general rule,
the riser on the hub is devised to have a capacity of between 35 and 50
per cent of the volume of the hub.

It is possible to produce gear blanks with a smaller proportion of feed
metal at the hub by the application of the total-reversal method of pour-
ing such as shown in Fig. 147. Other examples of the use of the reversal
method of applying hot metal to risers to obtain proper temperature
gradients in both metal and mold are shown in Figs. 76 and 124 and have
been explained under Gating.

Risers must be sufficiently numerous to feed all points where a tend-
ency toward unsoundness may exist. Figure 148 illustrates a casting

with many isolated heavy sections, necessitating a number of risers to take care of the shrinkage in all sections.

The placing of risers should not be left to the molder. A shop committee consisting of the foundry superintendent, the molding foreman, the patternmaker foreman, and the metallurgist should direct the placing of all gates and risers; and records of the effectiveness of the system should be kept for future reference. The use of models is an excellent way to study the proper positioning of gates and risers. Riser pads showing the proper size of risers to be used should be placed on all patterns by the pattern shop.

Fig. 148.—Riser requirements on a high-pressure, high-temperature casting.

Certain production castings readily lend themselves to the standardization of effective risers. For example, one steel foundry has prepared a table of the diameters and face dimensions of various sizes of solid gears to which are added columns giving the dimensions of the riser pads that apply. Such carefully prepared tabulations leave no room for guesses or indecision.

Blind Risers.—Blind risers are being used today in the production of steel castings to a much greater extent than at any time in the past. Blind risers were primarily used on large castings where riser applications were necessary at positions deep in the mold. Better understanding of blind risers has resulted from the use of gating into blind risers and from a knowlege of the effect of atmospheric pressure on the molten metal in the blind riser. Today many foundrymen are using blind risers at positions on small castings that nörmally might be considered positions for open risers.

Considerable study has been given to the functioning of blind risers by Taylor and Rominski[40] They illustrate the action of a blind riser in two steps. In Fig. 149 a casting is being fed by an open riser. This riser feeds the casting near its base in the manner that many blind risers are normally attached. As solidification and contraction proceed, atmospheric pressure alone is tending to move the liquid metal from the riser into the casting since gravitational force is balanced. With atmospheric pressure acting equally on all parts of the casting, the direction of effective action will be at the point offering least resistance. If solid steel is not
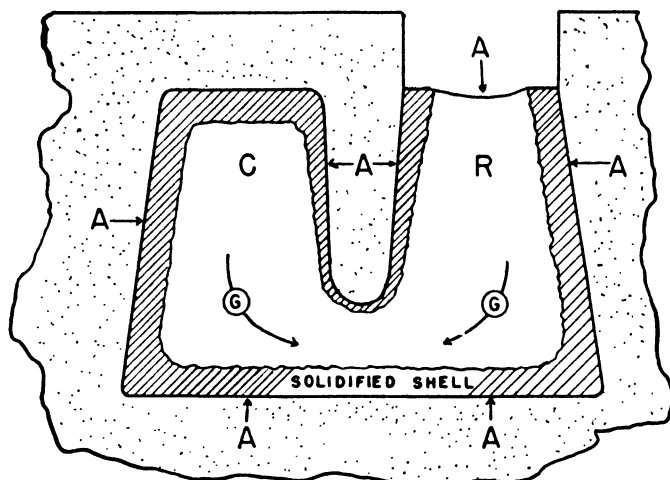
FIG. 149.—Atmospheric pressure acting without benefit of gravity to cause feeding.
(*Taylor and Rominski.*[40])

Legend: A = atmospheric pressure
G = metal head or force due to gravity
C = casting
R = riser
S = sprue

permitted to form on the surface of the open riser, or if it is weak enough to be ruptured at some point, the force will be at the surface of the metal in the open riser and will force the metal downward, then upward into the casting to feed that casting.

In Fig. 150 the casting is fed by a blind riser the height of which is less than that of the casting. This is a typical application of a blind riser. Prior to the solidification of the sprue, gravitational forces are similar to those of Fig. 149; but the sprue solidifies over, long before the casting solidifies, and in doing so it sets up a completely sealed system in which the liquid head in the casting is actually opposing the desired direction of metal feed. With very pronounced temperature gradients between the riser and the casting, it is possible to limit all the shrinkage in the riser.

In most cases a pronounced temperature gradient is not obtained; and since atmospheric pressure cannot gain entrance to the liquid metal, the system would work in reverse and the casting would then tend to feed the riser. If, however, the riser could be kept open at some point during its solidification, a point of entrance is provided for the atmospheric pressure to act on the liquid feed metal. If this force exceeds the gravitational force acting against it, the feed metal will flow from the riser and prevent the formation of shrinkage cavities in the casting.

Theoretically, it should be possible for a blind riser to feed a casting section to a height slightly greater than 4 ft. above it, if all conditions
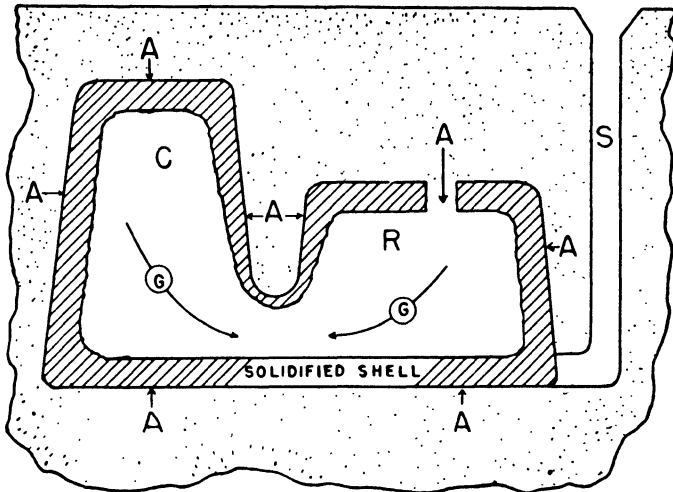
Fig. 150.—Metal head opposite to the direction of feeding. (*Taylor and Rominiski.*[40])

Legend: A = atmospheric pressure
G = metal head or force due to gravity
C = casting
R = riser
S = sprue

were perfect. Practically, however, the feeding conditions are far from perfect and feeding is limited to small sections which extend not much more than 2 ft. above the blind riser. In the manufacture of most small- and medium-sized castings for which blind risers are especially suited, it is seldom necessary that the casting section extend much above the top of the blind riser, inasmuch as blind risers are placed with their base on or near the parting line.

The manner in which atmospheric pressure acts to provide a sound casting when fed from a blind riser is given by Taylor and Rominski[40] as follows:

As soon as the mold is completely filled, the metal loses temperature rapidly to the sand and a skin of solid metal quickly forms at the mold-metal interface. As tem-

perature continues to drop, more and more metal solidifies and the liquid shrinkage accompanying the falling temperature tends to form a partial vacuum within the solidifying mass. The skin forming on all parts of the casting is impermeable except at the desired (open) spot on the riser. In this case as solidification continues in the casting proper, the atmospheric pressure plus any gas pressure in excess of this value acts like a piston on the metal in the blind riser, forcing it into the casting to feed shrinkage.

In other words, the system is functioning like a mercury barometer. Shrinkage is constantly tending to create the necessary partial vacuum in the casting, and atmospheric pressure, acting through the medium of the molten metal in the riser, is constantly relieving it. If solidification proceeds properly, with the parts most remote from the riser freezing first and progressing thence toward the riser, each successive amount of shrinkage is compensated by additional fluid metal forced in from the riser. The ingate, being small, freezes off first and completes the closed system.

It is to be noted that the direction of feeding is important. It is suggested that in controlling the direction of feeding through the establishment of temperature gradients, the gate should enter directly into the riser. If several blind risers are on a casting, then several ingates from a common sprue should be used.

It is possible for atmospheric pressure to produce defective castings unless its action is understood and care is exercised to prevent its undesirable operation. Atmospheric pressure can cause the caving or dishing in of the walls of a casting in unfed or incompletely fed sections. The collapsed walls of a casting section are the combined result of internal shrinkage, the formation of a partial vacuum, and the external action of atmospheric pressure. The caving takes place shortly after the casting section solidifies and while the strength of the steel is very low.

Holes and cavities may appear in castings at positions of sharp reentrant angles. The reentrant angles produce pronounced hot spots and weak points, permitting atmospheric pressure to break through at this point to act in a manner analogous to a blind riser. These break-through holes should not be confused with gas porosity. Differentiation can be made by the fact that a single hole or just a few exist at the reentrant angle; it will be observed upon exploration that they are continuous to the shrinkage cavity.

Methods of Applying Atmospheric-pressure Feeding.—Several methods can be used for keeping a blind riser open:

1. In cases where hot metal is used and an early feed demand develops in the casting, the riser may break through without any artificial means being provided. The uncertainty of such a break-through occurring makes the method undependable.

2. The method developed at the Dodge Steel Company[41] consists of embedding a round dried-sand core in the mold at the top of the blind riser, so that it extends into the riser well below the solid skin of frozen metal at the mold-metal interface and is in contact with the molten metal

in the casting in the manner shown in Fig. 151. The distance that the core extends into the riser is roughly equal to the radius of the riser. The core is a simple rod core of a type found in the average foundry except that a small venthole extends through the center. This venthole is not entirely necessary but is recommended. The only requirement is that the core be made sufficiently strong and refractory to withstand the pressure and temperature of the molten metal and permeable enough to allow passage of gas.

3. Any reentrant angle extending into the blind riser or adjacent to it will produce a hot spot at that point, owing to the inability of the molding material to conduct heat away as rapidly from this position as from the remaining surface of the blind riser. The rate of skin formation is
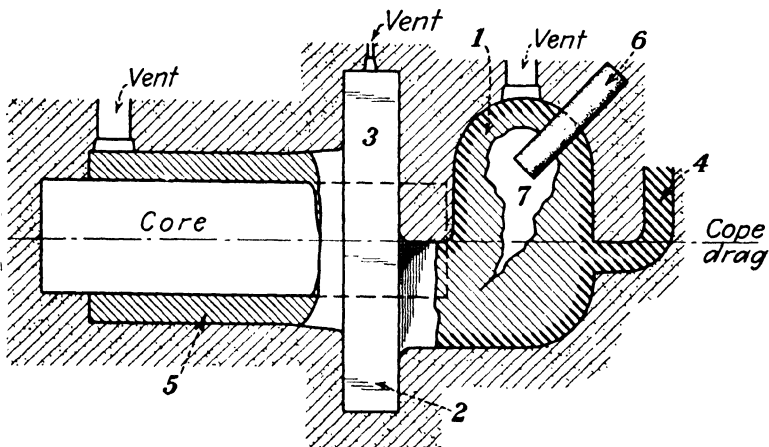
Fig. 151.—Williams core for blind risers.[41]

retarded at the hot spot. A partial vacuum is formed because of solidification shrinkage, and atmospheric pressure causes a break-through of the thin solidified skin into the hot spot. Such shapes as a sand wedge or cone, or a sand well, form hot spots. These shapes are a part of the mold and are not inserted as cores. A sharp-angle ingate entering the blind riser near the top of the riser will also provide a sharp reentrant angle and form a hot spot.

4. The use of thermit, either made up in the form of a core or placed at the top of the riser, has been found satisfactory.

5. Graphite rods inserted into the blind riser are used in place of sand cores. It has been established[42,43,44] that by the insertion of graphite rods into risers the solidification temperature of the metal in the risers is lowered and made more fluid and therefore more adaptable to filling the shrinkage cavity. The oxidation of the graphite within the confines of the riser cavity generates available heat that increases the temperature of

the riser metal and produces a more pronounced temperature gradient in the direction of feed demand. The graphite rod is inserted horizontally through the upper half of the riser. The ends of the rod are held in place by the sand mold. Atmospheric pressure may have access to the molten metal along the outside of the rod as well as through the rod.

6. A paper cup filled with powdered graphite and located on the top of the blind riser is used to provide an opening.

7. One method, developed at The Ohio Steel Foundry,[45] consists of placing a core on top of the blind riser to provide a sharp, thin edge of sand just on the top of the riser; this acts as a stress raiser and hot spot. The core is made of a carbonaceous mixture of sand plus 10 to 15 per cent powdered coal.

8. Another method developed at The Ohio Steel Foundry[46] consists of the use of a flat-disk core having a central portion in the form of a truncated cone. The core is set into the top wall of the blind-riser cavity and acts as a reentrant angle for hot-spot formation.

Advantages.—Blind risers have several potential advantages over open risers. They permit bottom gating into a riser, which results in a minimum of mold erosion, and at the same time compensate for the adverse temperature gradients established. They can frequently be more readily located in the mold with respect to the section to be fed. They can generally be designed for easier removal, which results in a reduced cleaning cost; and they afford a substantially increased yield as compared with open risers by conserving the total amount of molten metal required to produce the casting.

Disadvantages.—By placing the blind risers to the side of the sections that are to be fed, rather than above them, as in the case of open risers, a larger flask is necessary for producing a given job. Occasionally, blind risers are responsible for the trapping of dirt that would otherwise float out in the superimposed open risers.

Shape and Size.—The shape of blind risers for maximum feeding should be that of a sphere or a hemisphere superimposed upon a cylinder with the over-all height at least 1 in. greater than the diameter. If graphite rods or carbonaceous material is used for the introduction of the atmospheric pressure, the height of the riser should be such that a carbon pickup will not occur in the neck portion of the riser at the place where it is attached to the casting. It is desirable that the shrinkage cavity or the zone of increased carbon content in the riser should extend to, but not below, the upper level of the neck leading to the casting. This planning provides for normal variations in the temperature of the metal. If an estimate can be made of the volume of the last casting section to be fed, the volume of the blind riser should be fixed at approximately 10 per cent in excess of this amount.

The neck construction should be a square cross section with the upper edges rounded. An approximate 2-to-1 ratio of neck size to the thickness of the section to be fed is advisable for small castings. The ratio decreases inversely with the section thickness, but the neck should not be smaller than the section to which it is attached.[40] A neck-down core can, however, be used in the neck portion.

When several risers are used at different levels in the same casting, it is essential that a particular zone of feeding be relegated to each riser.

MOLD SET TO POUR AT AN ANGLE OF FIFTEEN DEGREES WITH "A" AT LOW END AND "B" AT HIGH POSITION. THEN MOLD IS ROTATED AFTER POURING WITH MOLD STANDING UP IN VERTICAL POSITION TO FLOOR AND WITH "A" IN TOP POSITION.

MOLD POURED IN REGULAR WAY.
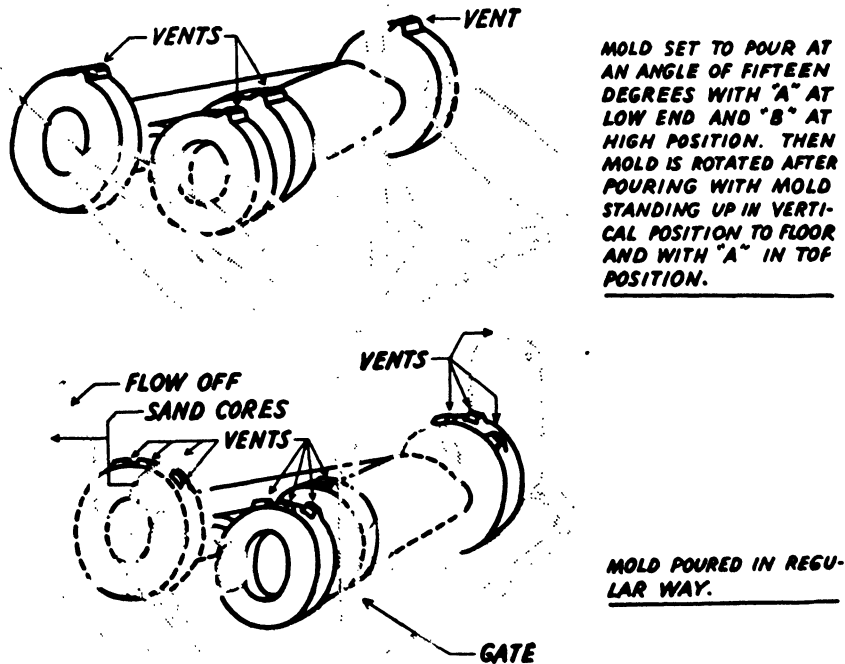
Fig. 152.—Blind-riser feeding substituted for the partial-reversal method of gating and feeding. (*Troy.*[47])

This can be done by placing external chills at desired positions of zone demarcation so that the section will freeze early, thus preventing one riser from feeding an area beyond the capacity of that riser. All blind risers should be vented to allow gases to escape the mold and to show when the mold is completely filled.

The application of blind risers with atmospheric-pressure cores, as a method of gating and feeding castings in preference to the partial-reversal method of gating and feeding, is receiving attention by some steel foundries. The illustration in Fig. 152 shows a typical application of blind risers.

Antipiping Compounds.—For any riser to function properly and at its maximum efficiency, atmospheric pressure must be allowed to act on the molten metal. Antipiping compounds are used to cover open risers for the purpose of insulating the metal and preventing it from freezing at the air-metal interface so that atmospheric pressure may act on the riser during the feeding of the casting. These compounds have been very effective in reducing the longitudinal extent of piping in a riser.

The effectiveness of antipiping compound is shown by Rassenfoss[39] in Fig. 153. No antipiping compound was applied to the sectioned riser

Fig. 153.—Section from casting and riser of 4- by 4- by 10-in. block: (a) no antipiping compound used; (b) granulated charcoal antipiping compound; (c) powdered carbon antipiping compound. (*Rassenfoss.*[39])

and casting illustrated in Fig. 153a. It will be noted that a skin has formed across the top of the riser, immobilizing a ½-in. layer of metal, which then served to prevent atmospheric pressure from acting on the liquid metal in the riser. As a result, the riser fed the casting inefficiently. Figure 153b shows a similar riser to which was applied a layer of granulated charcoal. The top of the riser has shrunk down nearly uniformly across its width. However, the macroetch reveals that a long, deep carbon pickup has taken place. Since carbon pickup is sometimes objectionable, especially if it extends into the casting proper—causing hard

spots that interfere with machining—the use of this particular antipiping compound is not much more efficient than none at all. In this case, the charcoal dissolved in the steel, producing a high-carbon steel of a lower freezing temperature, which filled the shrinkage cavity after the solidification of the low-carbon high-freezing temperature steel. In the case of Fig. 153c, powdered carbon was used as an antipiping material. Although some carbon pickup is noticeable, the size and extent of the high-carbon area are markedly less than in Fig. 153b. It is apparent from these illustrations that the riser efficiency can be truly measured only by considering the visible shrinkage area plus the area of chemical-composition change.

Antipiping materials to be effective must be on the riser as soon as possible after filling the riser. Rassenfoss[39] advises that the antipiping material be applied to the riser before the molten metal comes to complete rest in the riser. More efficient feeding will result since a minimum of heat is lost from the riser and the danger of partial skin formation on the riser is lessened. The antipiping material should cover the open riser to a depth of at least $\frac{1}{2}$ in. Foundrymen generally do not add a sufficient depth of antipiping compound to insulate the metal from the air. Instead of using antipiping compound, it has become the practice at some foundries to use graphite rods, run horizontally through open risers.[44] The level of the rod within the riser is from 2 to 3 in. below the surface of the mold.

The chemical compositions of different proprietary antipiping compounds differ considerably. Generally these compounds consist essentially of mixtures of carbonaceous matter and irreducible oxides, although some compounds are free of carbonaceous matter.

When placed on the surface of the molten steel, the carbon and any other elements present slowly oxidize, thereby generating heat. The nonmetallic matter in the compound remains as a residue, which serves to insulate the molten metal beneath. It is this combination of heat-generating and heat-insulating properties that serves to delay the freezing of the metal in the riser to such a marked degree.

Other methods rely on the use of a stream of oxygen across the metal or the use of an electric arc to keep the riser open. Such methods are practical only when used on risers larger than approximately 3 ft. in diameter.

Vents.—Vents are a necessary part of any system of gates and risers. The function of a vent is to permit the escape of gases from the mold cavity at such a rate as to avoid developing within the mold cavity any serious back pressure that would oppose the inflowing of metal. Vents also prevent the gases from becoming entrapped in the metal and forming gas cavities. Vents are not usually necessary in open, synthetic sands.

Experience has shown that round vents of sufficient diameter to evacuate mold gases at a proper rate will reveal frequently, when removed, a fine pipe in the casting. Vents of rectangular section are greatly to be preferred; and the thickness of such vents should not be greater than one-half the thickness of the section upon which they are imposed. Since the thickness of vents is prescribed by the casting section, their width has to be determined in relation to the devised pouring rate to ensure that serious back pressure is not generated during pouring. If the vent is small, such as ¼ in. in diameter, it may be of a circular cross section.

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CHAPTER VIII

CONTRACTION IN THE SOLID STATE

Contraction in the solid state is briefly reviewed in Chap. V in order to discuss the total contraction resulting from the cooling of liquid steel to atmospheric temperatures. This contraction is represented in Fig. 42.

Free Contraction.—The first measurements of the solid contraction of cast steel were made by Wüst and Schitzkowski[1] on a cylindrical test bar. They found that the solid contraction decreased with increase of carbon content from 2.39 per cent for 0.11 per cent carbon steel, to 2.22 per cent for 0.90 per cent carbon steel. Körber and Schitzkowski[2] measured the solid contraction of a 0.28 per cent carbon cast steel. Both acid and basic open-hearth steels were studied. Bars of different diameters were cast horizontally. The results on the freely contracting bars indicated that the contraction (2.18 per cent) was independent of the diameter of the bar or the method of producing the steel.

A series of extensive and accurate measurements on the effect of composition on the solid contraction of cast steels were made by Briggs and Gezelius.[3,4] The type of curves obtained is illustrated in Fig. 44. Solid contraction decreases with an increase of the carbon content of the steel (Table XLIV), and alloy additions do not have a marked effect upon the total amount of contraction (Table XLV). The following summary of results shows that there is farily good agreement on measurements of solid contraction:

Authorities	Carbon, per cent	Solid contraction, percentage of length of solid at solidification point
Expansion laboratory tests[5].	Nil–0.8	Decreases from 2.32 to 2.15
Wüst and Schitzkowski.....	0.1–0.9	Decreases from 2.39 to 2.22
Körber and Schitzkowski...	0.28	Decreases from 2.18
Briggs and Gezelius........	0.1–0.9	Decreases from 2.47 to 2.18

Hindered Contraction.—The free contraction of cast steel does not portray the actual conditions encountered in the contraction of a commercial steel casting. The steel-casting design, flask construction, sand conditions, and mold inserts are additive factors in preventing a casting from contracting freely. This fact may be easily seen by comparing the

values for a freely contracting casting with the patternmaker's shrinkage of $\frac{3}{16}$ in. per ft. (1.56 per cent), normally used for steel castings.

Hindered contraction of cast steel was studied by Körber and Schitzkowski[2] by the use of insert bars in the mold and by the use of bars with
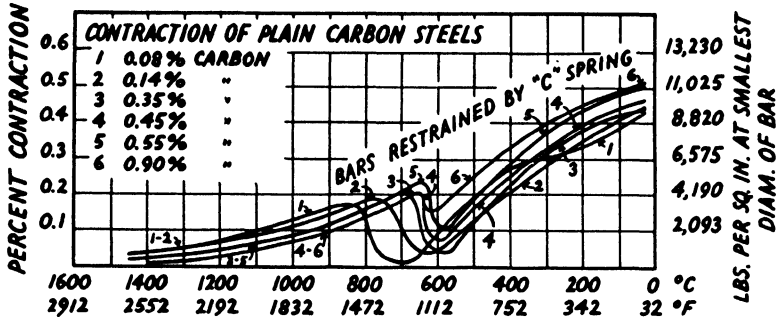
FIG. 154.—Contraction of carbon cast steels. Restrained by a heavy tension spring *C*. (*Briggs and Gezelius.*[3])
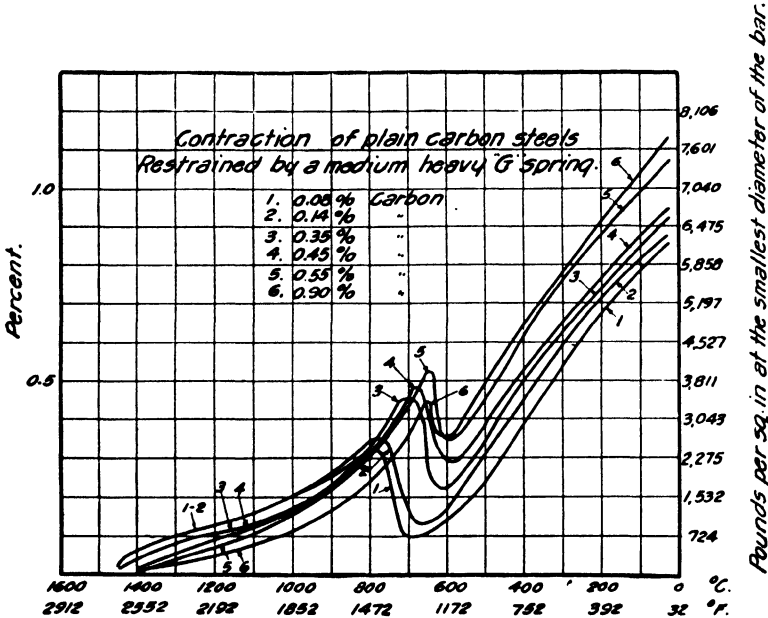
FIG. 155.—Contraction of carbon cast steels. Restrained by a medium tension spring *G*.

flanges. It was concluded that hindered contraction was dependent upon the nature of the mold resistance, and that hindered contraction could cause steels to crack. Briggs and Gezelius[3,4] evaluated the amount of stress that would arise from hindered contraction. Information was collected by experimentation on cast-steel bars that had cooled uniformly

throughout their length from the casting temperature. The contraction was hindered by means of calibrated springs. The results are presented in Tables XLIV and XLV and illustrated in Figs. 154 to 157.

It will be noted that for carbon steels the hindered bars contracted similarly to the freely contracting, for at any given temperature the amount of contraction is greater with lower carbon contents. In passing through the critical range, the order was reversed so that at the lower
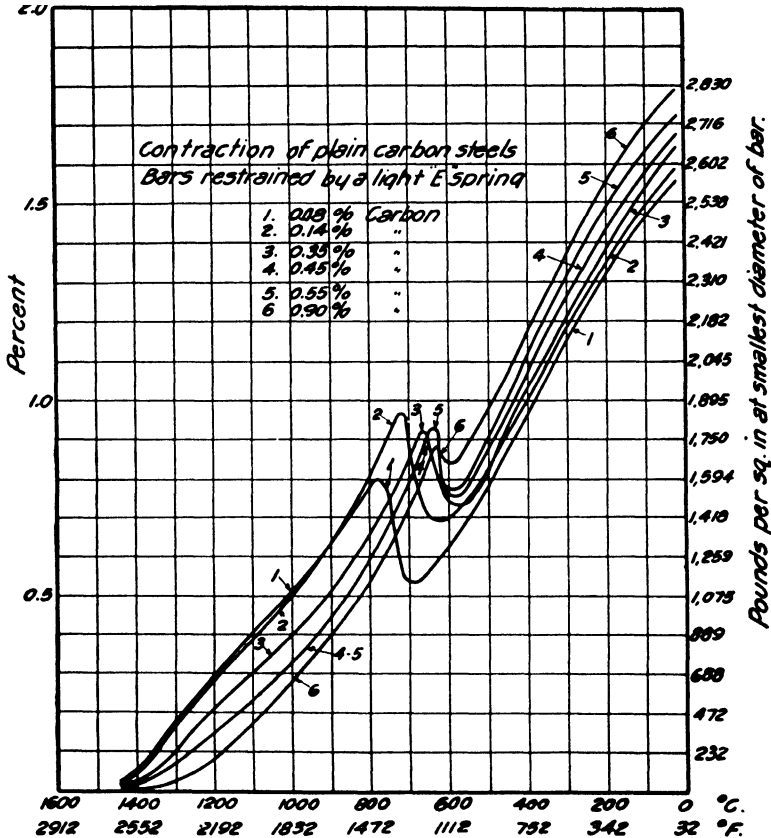
Fig. 156.—Contraction of carbon cast steels. Restrained by a light tension spring *E*.

temperatures the amount of contraction in the hindered bars was greater with higher carbon steels.

Carbon cast steels under hindered contraction show greater expansion during the critical range than that exhibited by the freely contracting bars. This increase in expansion is greater when the lightest spring is used. Also, the percentage of the total contraction occurring before the critical range is reached decreases as the tension restraining the bar increases. The amount of contraction before and after the critical range,

although different for each spring, is not greatly affected by the carbon content. The amount of expansion passing through the critical range decreases as the carbon content increases.

For the alloy cast steels, the amount of contraction taking place in both freely contracting and hindered-contracting bars prior to reaching the critical range is, in general, dependent upon the temperature at which the critical range occurs. The total amount of hindered contraction obtained corresponds closely to that exhibited by carbon cast steel. The
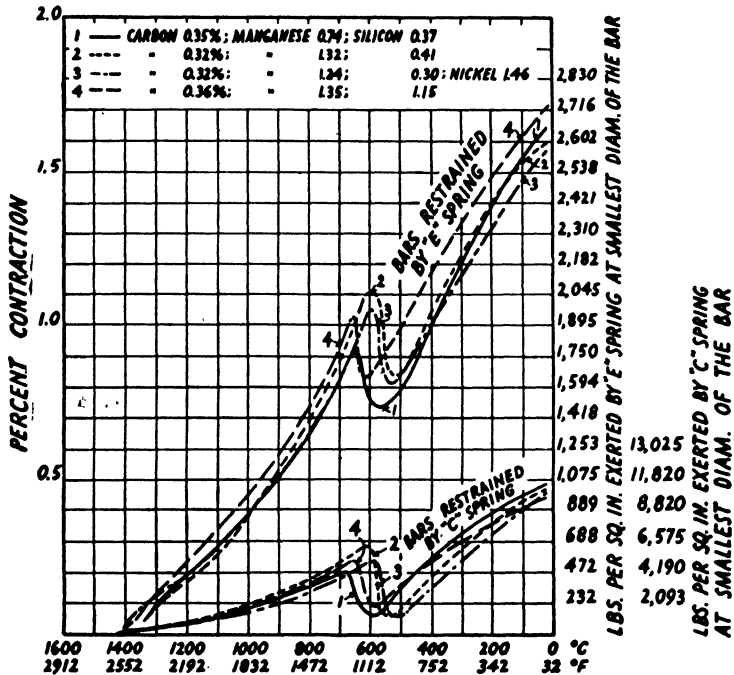
Fig. 157.—Contraction of alloy cast steels. Restrained by heavy (*C*) and light (*E*) tension springs. (*Briggs and Gezelius.*[4])

total contraction of the different alloys varies somewhat with different amounts of tension on the bars. The amount of expansion shown by the alloys during the critical range varies considerably.

The data obtained with the light-tension *E* spring probably represent more closely the stresses encountered by the average commercial casting, since the total contraction recorded under this tension approximates the patternmaker's shrinkage of $\frac{3}{16}$ in. per ft. (1.56 per cent). Values of 1.58 to 1.78 were found for cast steels in the commercial ranges. Cast steels showing a total contraction of 1.58 to 1.78, after being cooled to room temperature, will be under a stress of 2,600 to 2,800 p.s.i. because of hindered contraction.

TABLE XLIV.—CONTRACTION OF CARBON-CAST-STEEL BARS RESTRAINED BY SPRINGS

Briggs and Gezelius[3]

Carbon, per cent	Contraction above critical range, per cent	Contraction within critical range, per cent	Contraction below critical range, per cent	Total contraction, per cent
Strong-tension spring (C)				
0.08	0.15	0.14	0.45	0.46
0.14	0.17	0.13	0.44	0.48
0.35	0.20	0.15	0.42	0.47
0.45	0.21	0.14	0.41	0.48
0.55	0.24	0.13	0.40	0.51
0.90	0.20	0.06	0.38	0.52
Medium-tension spring (G)				
0.08	0.32	0.22	0.76	0.86
0.14	0.35	0.22	0.75	0.88
0.35	0.46	0.24	0.70	0.92
0.45	0.49	0.20	0.66	0.95
0.55	0.52	0.18	0.73	1.07
0.90	0.44	0.09	0.77	1.12
Light-tension spring (E)				
0.08	0.80	0.27	1.03	1.56
0.14	0.96	0.27	0.90	1.59
0.35	0.92	0.18	0.90	1.64
0.45	0.90	0.13	0.90	1.67
0.55	0.92	0.15	0.95	1.72
0.90	0.88	0.04	0.95	1.79

The data recorded with the medium-tension *G* spring probably approximate conditions existing in castings that, owing to design or to unusually high mold resistance, are unable to contract to any great extent. These conditions are not infrequent, especially in long-cored-construction castings. Total contraction values of 0.90 to 1.05 per cent are recorded for commercial-grade cast steels. This amount of contraction is equivalent to a stress of 6,500 to 7,300 p.s.i. acting on the casting while the casting is under the hindered-contraction conditions.

The tensions exerted by the strong *C* spring are probably greater than any encountered in commercial practice. If they do form, the casting will tear to relieve the stresses. It was pointed out that under the influence of the tension stresses, the amount of expansion resulting while the casting passes through the critical range is greater than that recorded in a freely contracting bar (see Figs. 154 to 157). This point is significant, since in a casting of various sections it is possible that one section may be

TABLE XLV.—CONTRACTION OF ALLOY-CAST-STEEL BARS RESTRAINED BY SPRINGS
Briggs and Gezelius[4]

Type of steel and per cent of alloy*	Contraction above critical range, per cent	Contraction within critical range, per cent	Contraction below critical range, per cent	Total contraction, per cent
Strong-tension spring (C)				
Carbon (0.35)	0.20	0.15	0.42	0.47
Manganese (1.32)	0.28	0.22	0.39	0.46
Nickel (3.00)	0.24	0.18	0.35	0.40
Chromium (1.03)	0.21	0.17	0.43	0.47
Copper (1.39)	0.25	0.17	0.41	0.48
Molybdenum (0.39)	0.31	0.12	0.30	0.49
Vanadium (0.25)	0.21	0.12	0.35	0.44
Mn-Si (1.35)(1.15)	0.24	0.16	0.35	0.43
Ni-Mn (1.46)(1.24)	0.23	0.18	0.38	0.44
Mn-Mo (1.41)(0.37)	0.26	0.09	0.28	0.45
Mn-V (1.41)(0.16)	0.24	0.16	0.45	0.53
Ni-Cr (2.88)(0.91)	0.37	0.17	0.17	0.37
Medium-tension spring (G)				
Carbon	0.46	0.24	0.71	0.92
Manganese	0.60	0.33	0.64	0.90
Nickel	0.56	0.28	0.61	0.89
Chromium	0.41	0.21	0.78	0.99
Copper	0.49	0.21	0.73	1.01
Molybdenum	0.67	0.14	0.52	1.05
Vanadium	0.45	0.11	0.64	0.98
Mn-Si	0.56	0.22	0.70	1.01
Ni-Mn	0.55	0.28	0.58	0.86
Mn-Mo	0.64	0.10	0.50	1.03
Mn-V	0.55	0.24	0.69	1.00
Ni-Cr	0.87	0.20	0.21	0.88
Light-tension spring (E)				
Carbon	0.92	0.18	0.90	1.64
Manganese	1.12	0.27	0.73	1.58
Nickel	1.13	0.25	0.75	1.64
Chromium	0.93	0.24	0.97	1.65
Copper	0.99	0.19	0.90	1.70
Molybdenum	1.28	0.10	0.61	1.78
Vanadium	0.90	0.15	0.93	1.67
Mn-Si	1.03	0.20	0.88	1.72
Ni-Mn	1.06	0.26	0.76	1.58
Mn-Mo	1.25	0.09	0.57	1.72
Mn-V	1.08	0.22	0.85	1.72
Ni-Cr	1.48	0.14	0.26	1.60

* Carbon kept between 0.32 and 0.37 per cent, manganese 0.55 to 0.80 per cent, and silicon 0.25 to 0.40 per cent, unless otherwise specified.

expanding while another section is contracting because of temperature variations and gradients. This leads to the formation of important stresses, which in turn increase, since under stress-active systems the amount of expansion becomes greater.

An interesting circumstance of practical importance, which throws some light on the mechanics of crack formation, is shown by the data obtained on the hindered-contraction bars. In the case of light hindered contraction, where conditions are similar to usual casting conditions, it will be noted that at 1300°C. (2372°F.), the temperature range most susceptible to cracking, the 0.08 per cent carbon bar has contracted 0.18 per cent and has developed a stress of 425 p.s.i., whereas the 0.90 per cent carbon steel has contracted only 0.002 per cent with a corresponding stress of about 50 p.s.i.

This effect is even more noticeable in the case of bars under high hindered contraction. For example, the 0.14 per cent carbon steel at 1300°C. (2372°F.) has contracted 0.006 per cent, which results in a stress of 1,200 p.s.i. At the same temperature, the 0.35 per cent carbon steel has contracted only 0.003 per cent and thus has to resist a stress of only 600 p.s.i. However, the stress magnitude is not the only matter of consideration in this case, because it will be subsequently shown from Chap. IX that the lower carbon steel has the greater strength and ductility at the very high temperatures.

It should, of course, be realized that a study of contraction stresses operative at the various high temperatures has no bearing on the ultimate strength of the steel, except that it shows that the slower contracting steels may have a lower yield strength at the higher temperatures, since the bar in contracting could develop a stress only equal to its yield strength. When the yield strength is exceeded, the steel yields; and upon further cooling—and thus further contraction of the steel—the stress is again built up to the new yield strength owing to the fact that the strength properties increase at progressively lower temperatures.

From a study of the data it seems probable, however, that the true yield strength of the steels is not recorded, since the stresses upon the bars at any one temperature vary considerably with the type of contraction. For example, the manganese-nickel steel of Fig. 157 at 1200°C., when under low hindered contraction, shows a stress of 472 p.s.i., whereas under high hindered contraction a stress of 885 p.s.i. is recorded. This would indicate that the rate at which the stress is applied has considerable to do with the amount of stress present at any one temperature under different types of loading. In the case of a low hindered contraction it seems possible that the yield strength is not reached even at the higher temperatures.

Contraction Stresses.—Prior to 1925, very little information was available on the mechanics of the formulation of casting stresses. Pub-

lications consisted of narrative treatments of the subject devoted to general descriptions of observed conditions. In 1927, Malzacher[6] discussed certain fundamentals concerning the formation of stresses in steel castings. He suggested that stresses originate from (1) mold resistance and (2) variations of metal thickness, establishing thermal gradients, with the total casting stresses the sum of (1) plus (2). This indicates that internal stresses occur because of sudden changes in the form of the casting, which produce changes in the direction of contraction and differences in the size of the sections of castings, thus giving rise to different rates of cooling.

The work of Körber and Schitzkowski, followed by that of Briggs and Gezelius, definitely points out that the primary source of internal stresses in castings is contraction stresses. It is therefore essential that the underlying principles involving contraction stresses be stated. These may be listed as follows:

1. There are no contraction stresses (acting linearly) in a freely contracting cast bar.

2. In a stressed cast bar the resulting stresses are independent of the length of the bar.

3. Stress magnitudes vary inversely with the cross section in which they appear.

4. Internal stresses in a casting are proportional to the modulus of elasticity, to the coefficient of contraction, and to the differences of temperature involved.

The above points can be developed mathematically.[7] As long as tension and compression stresses acting along the casting produce only elastic deformation, the stresses will be temporary; and although the relations may become somewhat involved, they can be attacked clearly from the mathematical viewpoint. Complications arise, however, for a stress system is usually rather complex, since other than linear tension and compression stresses, such as circumferential, radial, and shear stresses, may be developed in the cast member. To make it more complex, when permanent deformation exists, some internal stresses are left in the metal. To exceed the elastic limit of a metal is not very difficult, since metal at temperatures approaching melting has exceptionally low elastic limits and the deformation taking place is principally plastic. It is well known that stresses causing initial yielding at low temperatures decrease about linearly with increasing temperature but become nearly independent of the temperature when the melting point is approached. Thus metal crystals have a small but finite elastic limit at temperatures approaching the melting point.

In the discussion of stresses in castings, it should be remembered that besides the stresses developed because of cross-sectional differences, originating in temperature variations in the metal, there are mold conditions

that can cause stresses. In this case, however, the stresses arise from the prevention of normal contraction by the rigidity of sand molds and cores. A combination of stresses resulting from differential cooling of metal sections and from the action of the mold-hindering conditions may be present simultaneously within the casting. The magnitude of this stress depends upon the rate of the stress application, or, to state it in another way, the stress depends upon the rate of plastic deformation.

$$S = K \frac{dp}{dt}$$

where S represents stress, p is the plastic deformation or elongation or, if one prefers, the hindered contraction, and t is the time. The magnitude of p therefore depends upon the amount of stress acting on the casting and the length of time that it is effective. Plastic flow of a metal occurs only as long as there is a force acting upon it, and the rate at which it takes place depends upon the magnitude of the force.

Thus, when metal is cast under conditions that tend toward high resistance to contraction, the magnitude of the stresses acting upon the casting is high, the rate at which they build up is fast, and the amount of plastic deformation that takes place is great. If, however, the resistance to contraction is not high, then the other factors bearing on it will be correspondingly lower.

This relates to plastic deformation arising from contraction stresses. It should be added that if at any time the cooling cycle of the stressed casting could be arrested at some high temperature, the magnitude of the stress would be progressively less with time, since plastic deformation of the metal would reduce the stress to zero. From this it can be seen that the cooling rate of the metal is important in considering the magnitude of the stress. In two hindered-contracting castings, acting under like conditions except for the cooling rate, the casting that cools the more rapidly will build up the higher stresses.

Stress Concentration.—All abrupt changes in section, sharp corners, casting irregularities, and hot spots caused by temperature gradients are potential positions of stress concentration. When stresses arising from hindered contraction of a large casting concentrate at one point, it is easy to see that stresses may assume high values. It is because of this that many foundrymen have been skeptical that contraction stresses arising in a casting are independent of the casting length, since they are able to point to the fact that many long castings fail while short castings do not. This failure in long castings is not the result of greater stresses developed because of their length but is due to the fact that these stresses may centralize in one place because of hot spots, sharp corners, etc., and the

combined stresses, centralized in a long casting, will be much greater than those centralized in a short casting.

It will be observed that stresses are important in the following manner: (1) the stresses that occur at high temperature where tearing takes place, and (2) the stresses that are residual in the casting at room temperature. Probably the most serious defects with which the steel-casting manufacturer contends are cracks, and of these the hot-tear crack is the most prevalent. The subject is of such consequence that the entire next chapter has been devoted to it.

Magnitude of Stresses at Room Temperature.—The magnitude of stresses generated in steel castings under hindered contraction has been studied by Briggs and Gezelius.[3,4]   If a casting contracts only 1 per cent instead of the free contraction of 2.4 per cent, then a stress of approximately 7,000 p.s.i. will act on the casting because of hindered contraction
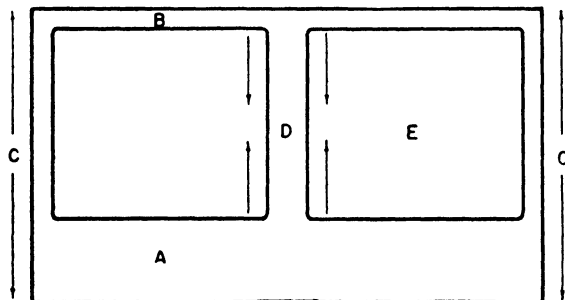
FIG. 158.—Hindered contraction conditions in a casting.   (*Briggs.*[7])

alone. If, however, the casting is removed from the mold, hindered contraction may no longer be effective, and the stress caused by hindered contraction may be removed.

For example, a flanged bar cooled in a hard mold to room temperature without cracking, contracting only 1 per cent, will generate a stress of 7,000 p.s.i. acting along the bar because the mold exerts a force against the inside faces of the flanges. If the mold is removed, the force is removed, and the casting is no longer under stress from hindered contraction resulting from the mold. This does not mean that castings are free from stress when removed from the mold. Hindered-contraction stresses arise also from the shape of the casting. For example, in the casting illustrated in Fig. 158, cores *E* and *E'* were very collapsible, and stresses were not formed by hindered contraction caused by mold resistance. It is quite possible, however, that a tear will be found when *D* joins *A*. In this case, stresses are built up owing to (1) variation in metal section and (2) hindered contraction resulting from the design and shape of the casting, where one member hinders the free contraction of another member.

If this casting is removed from the mold, the stress resulting from hindered contraction caused by casting design is still effective.

Variations in metal section may cause castings to be under stress, the magnitude of which may mount as high as the ultimate tensile strength of the steel.   These high stresses are developed by unequal cooling of the metal sections.   One section may be expanding while cooling through the critical range, and at the same time another section at a different tem-
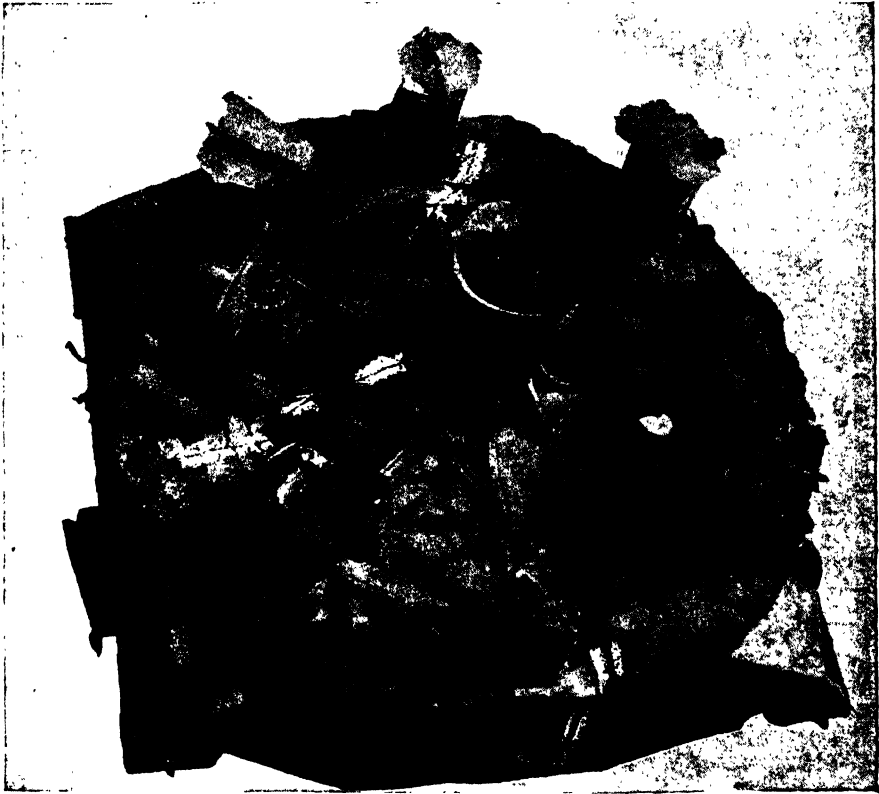
Fig. 159.—Cold cracks in a steel casting.

perature may be contracting.   This condition may result in a difference in movement of as much as 0.3 per cent.   Since these sections are integral, movement is not possible without producing stresses of large magnitude within the casting.   If 30,000 p.s.i. were assumed as the elastic limit, with a modulus of elasticity at 30,000,000 lbs., then a resistance to a movement of 0.1 per cent would be equivalent to a stress equal to the elastic limits at room temperature.   This would be somewhat different at temperatures in the critical range but may be taken as a rough indication of the stress magnitude that may exist.

Cracks.—Cold cracks are the result of large inherent contraction stresses produced during cooling from the solidification temperature to room temperature. These cracks appear at comparatively low temperatures, usually from freezing temperatures to about 500°F.

Cold cracks are continuous and usually exist as a single-crack system with but few branching cracks. The crack is not jagged or irregular, as is a hot tear. Cold cracks always appear on the surface and may or may not extend entirely through a section. An example of a cold crack caused by high contraction stresses is shown in Fig. 159.

Cold cracking is prevalent in large, complex-shaped castings. It has been the experience of large casting manufacturers that castings on the foundry floor may crack if the room temperature drops considerably. The added stresses resulting from the further contraction of the casting from 70 to 32°F. were sufficient to exceed the ultimate strength of the casting and produce cracking.

In order to prevent cold cracking, castings should be cooled as uniformly and slowly as possible; or they should be removed from the mold at temperatures about or above 500°F. and placed in a stress-relieving furnace, the temperature of which is approximately that of the casting.

Distortion.—Contraction stresses may not be high enough in a casting to produce tears or cracks, but they may be sufficient to distort the shape of the casting. Hindered contraction from either mold design or casting design or both is primarily responsible for distortion. These stresses produce plastic deformation in the casting members. The casting member does not contract in accordance with the patternmaker's calculated shrinkage; hence the section is out of line, or distorted.

In some cases it is necessary to use two or three shrinkage allowances in the construction of a pattern. Experience has dictated that in order to make ring castings, for example, it is advisable to construct the pattern out of round, so that the unequal contraction on cooling will produce a casting of the circular requirements desired. Distortion may be eliminated or materially reduced by preventing hindered contraction caused by molds and casting design.

Factors Contributing to Hindered Contraction, and Methods of Prevention.—There are two major factors that result in the hindered contraction of steel castings:

1. Mold and core resistance
2. Casting design

Mold and Core Resistance.—Mold resistance is effective in preventing contraction. This is not only because of the fact that some sands have a high compression strength and resist the contraction; the mold hardness and the amount of sand around the casting also are important. For example, the sand between H-shaped sections, such as encountered in

pulley wheels, gear blanks, valve bodies, etc., restricts the casting from contracting. Furthermore, the sand mold outside this area is rigidly shaped; hence it is not only a question of overcoming the compression strength of the sand between the H sections but also one of overcoming the resistance of the entire mold to the casting movement.

It is common knowledge that green-sand molds give less mold resistance than dry-sand molds, for dry-sand molds have greater compressive strength at room temperature and at some elevated temperatures. The character of the strength of mold and cores should be judged at elevated temperatures, since it is at these temperatures that hindered contraction is most effective in promoting high stresses and casting failure. Sand-mold and core mixtures change their strengths considerably at the various temperatures. A sand that may test only 6 p.s.i. compression strength at room-temperature conditions may develop strengths of upward to 1,000 p.s.i. at high temperatures. The nature and the effect of molten-steel temperatures on sands and cores in resisting contraction are more fully noted in Chap. XI.

Mold and core resistance can be relieved partially or nearly completely by the employing of friable or collapsible molds and cores, relieving blocks, arbors and sand mixtures that have low strengths at all elevated temperatures. Molds and cores may be so constructed that they consist of a thin outer wall of sand at positions in contact with the metal, while the inner portion of the core is filled with charcoal, loose sand, sawdust, hay rope, cinders, or even nothing at all. Such materials have little or no strength at elevated temperatures and offer no resistance to the contraction of the casting. The disadvantage in their use is that these materials contaminate the sand for further use.

The use of relieving blocks is excellent for reducing hindered contraction. Castings that have superimposed risers or flanges, which normally would cause tearing or distortion of a casting, can be made ineffective by molding in blocks of wood next to them in the direction of casting contraction. These blocks are separated from the casting protuberances by a thin layer of sand. Shortly after the casting has been poured and at about the time it completely solidifies, these blocks are removed from the mold by hand or by a crane. The open spaces will offer no resistance to the contraction of the casting. In large molds, such as in the production of locomotive beds, eight or ten large relieving blocks are used as standard practice.

The use of mold-relieving arbors is popular in the manufacture of castings of certain designs. In most cases a relieving arbor is used in connection with large castings, but occasionally it is used for castings of only a few hundred pounds. An example[8] of the use of an arbor is cited in connection with the production of a two-armed strut casting such as is

shown in Figs. 160 and 161. Five to 8 min. after pouring, the bars holding the drawbacks in the mold are loosened to permit the casting to move with the least possible mold resistance. Twenty minutes after

FIG. 160.—Two-arm strut casting. (*Duma and Brinson.*[8])
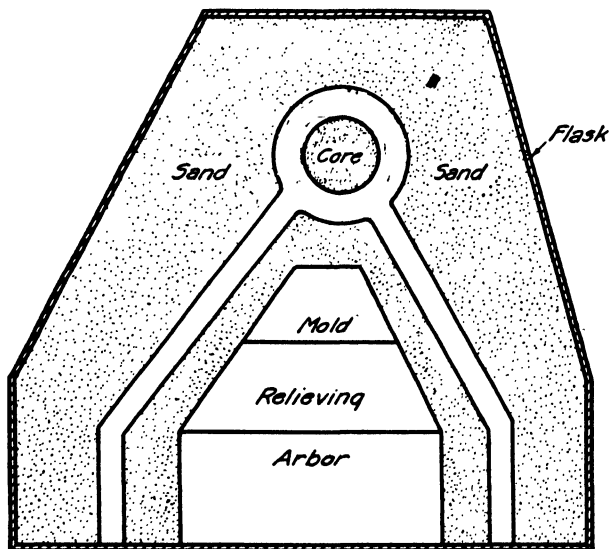
FIG. 161.—Mold-relieving arbor.

pouring, all mechanical restraint on the drawbacks is removed. A readily detachable green-sand drawback is used in the crotch to keep resistance low.

Half-cylindrical cores, such as those used in turbine casings and similar castings, can be made collapsible by constructing a thin exterior sand shell on a cast-iron arbor so that the center of the core is hollow. The arbor is constructed so that there are only three of four main crossbars in it. These bars can be reached from the cope surface of the mold. After the casting has solidified, the main crossbars are broken with a sledge, allowing the casting to contract with but little hindered contraction.

Mold resistance can arise from the negligent use of core and mold irons or rods. Utmost care should be exercised to see that core irons are straight and pointed on the ends, so that they will not interfere with the collapsibility of the core. Considerable care is given by some steel foundries to the placement of all the rods and reinforcing bars in such a manner that they protrude from the ends of the flask, so as not to hinder contraction. Studies of the strength of sand at elevated temperatures will permit the selection of weak sand mixes that have sufficient strength at atmospheric temperature to permit their use as molding or core sands.

Hindered Contraction from Casting Design.—Steel castings should be so designed that there will not be closed, stress-active systems where temperature gradients and hindered contraction may lead to high stresses. Casting members that resist other parts of the casting from contracting should be studied, in order to ascertain whether changes in the manner in which these sections join can be made. Abrupt changes in section should be modified to smooth, streamline, sloping changes. Small-section tie members should not be cast in but welded in after the casting is made. Greater use of cast-weld construction should be made (see Chap. XV) to reduce hindered contraction.

Assistance on casting design should be given by the metallurgist to the design engineer so that he may use more uniform sections and thus eliminate unnecessary hot spots, since these points are positions of centralization of stresses. The judicious use of external chills can do much toward eliminating hot spots; the application of ribs, brackets, and cracking strips serves to check hot-tear formations. The function of brackets, etc., is to solidify earlier than the sections to which they are attached and thus to act as a bond in preventing hot tearing. Brackets do not eliminate contraction stresses; they merely tend to prevent the stresses from centralizing at one point and rupturing the casting. The strength added to the casting skin by the presence of brackets acts to spread converging stresses over a wide area.

Even a generous supply of brackets will be of no use on some castings; though the ribs themselves remain solid, cracks occur between them, and sometimes even the brackets crack. In such cases the designer must come to the aid of the foundryman in providing gradual transitions between joining sections. In the application of brackets, care should be

exercised to see that they are cut in beyond the affected area.   If these steps are not taken, cracking may be found at the positions where the continuity of brackets ends.

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CHAPTER IX

HOT-TEAR FORMATION
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In the manufacture of steel castings, defects in the form of cracks are occasionally found. These cracks have a very irregular and jagged appearance; and since the fracture face exhibits oxidation and shows a heat effect, these cracks are known as "hot tears." From various practical considerations and from certain experimental observations it has been established that hot tears form at temperatures not greatly below the solidifying temperature of steel.

Types of Hot Tear.—Hot tears have been divided into two groups:[1]

1. External hot tears
2. Internal hot tears

The external hot tears are so called because the crack can be plainly seen on the surface of the casting. The crack first opens on the surface of the casting and proceeds inward. Such a crack has the appearance of a tear; i.e., the face of the crack is ragged. The crack is wider at the surface, is uninterrupted in depth, and in some cases may traverse the entire section (Figs. 162a and b). The cracking or the failure of such a casting could be considered comparable to the failure of a test specimen under load at a high temperature; the cracking begins at surface irregularities, sharp corners, changes in section, and the like, which serve as points of stress concentration. The stresses active in their formation appear to be tensile or shear stresses.

The internal hot tears (Figs. 163 and 164) are found near the center of the section, or the portion of the section that solidified last. They are very ragged, usually with numerous branches. They have no definite line of continuity and nearly always exist in groups. These tears seldom terminate at the surface. When they do appear at the surface, they are usually very small and difficult to locate. Upon exploring downward, they will be found to branch out and become more pronounced.

The internal hot tears are regarded as the most dangerous type, since they may be present in a section the surface of which is without defect. Their presence is not realized until they are uncovered by machining or disclosed by radiographic examination. It is well substantiated that internal hot tears present in a casting in service may extend until they traverse the section. This usually renders the casting unserviceable, as, for example, in the case of a pressure casting.
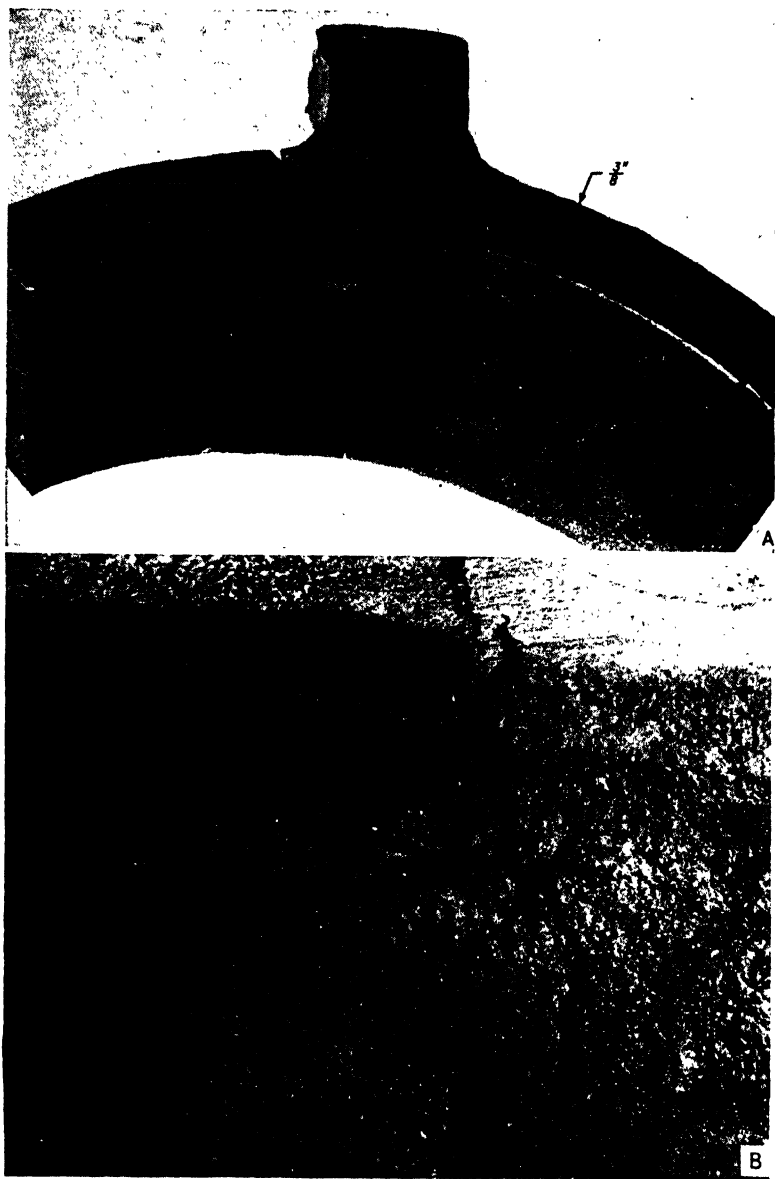
FIG 162.—External hot tear: (*a*) in a small casting; (*b*) in a large casting (⅔ actual size).

Fig. 163.—Radiograph of internal hot tear in a 2-in. section of a steel casting. (*Courtesy of U.S. Naval Research Laboratory.*)
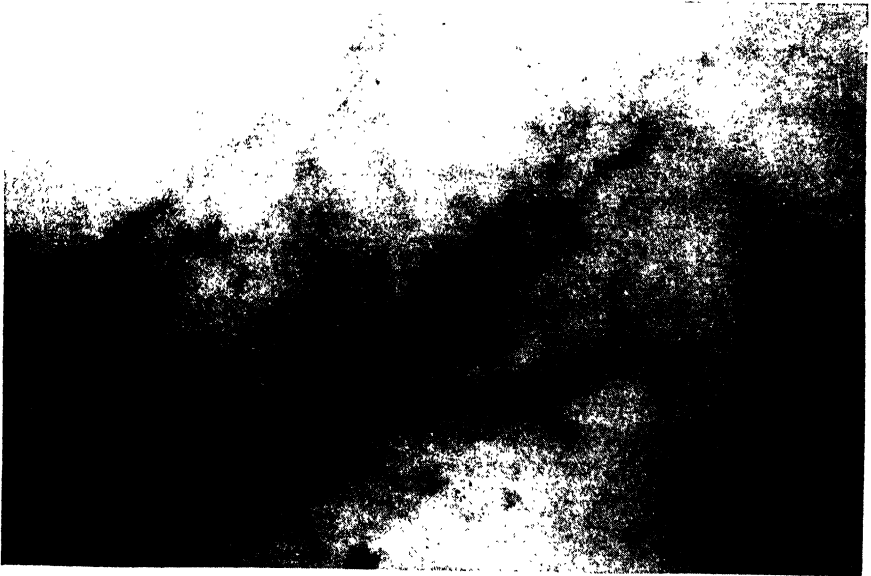
Fig. 164.—Radiograph of internal hot tear. Tear in section is aproximately 4 in. thick and extends about 3 ft. without coming to either surface of the steel casting. (*Courtesy of U.S. Naval Research Laboratory.*)

Occurrence of External Hot Tears.—Hot tears occur because the casting or a portion of the casting is under stress of sufficient magnitude to cause cracking to take place. This stress originates in the design of the casting, the manner in which the casting cools, and the construction of the mold. Stresses arising from any one of these conditions may be sufficient to produce hot tears.

Internal stresses in a casting are proportional to the modulus of elasticity of the cast steel, to its coefficient of contraction, and to differences in temperature within the casting. Since the modulus of elasticity decreases with increasing temperature, the magnitude of the stresses necessary to cause permanent deformation and permanent stresses in the casting also decreases. Steel has a relatively high coefficient of contraction, and the temperature gradients present in a casting may be high. Accordingly, the occurrence of high stresses is frequent.
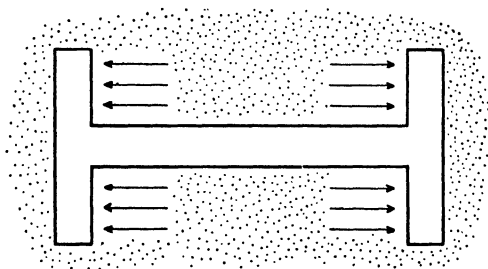
Fig. 165.—Bar with flanges.

As steel cools from its solidification temperature to room temperature, it contracts. If it is allowed to contract freely, it will have a volumetric contraction of 7.2 per cent for a 0.30 per cent carbon cast steel, which is a linear contraction of a 2.4 per cent.[2] Steel castings, however, do not contract this amount, owing to their shape and the fact that the mold material in which they are cast acts to restrain contraction. The normal patternmaker's shrinkage of 1.87 per cent, which is used in the construction of patterns, is in itself a value of hindered contraction. Some castings may be produced with only a total linear contraction of 1 per cent or even less.

It has been established (Figs. 155 and 156) that as the total linear contraction in the solid state is reduced by hindrance, the stresses acting on the casting increase.[2] If, for comparison, it is assumed that the amount of stress at 2000°F. is 875 p.s.i. for a total contraction of 1.65 per cent, then at 0.95 per cent contraction the stress would be approximately 1,050 p.s.i. In other words, hindered contraction increases the stress acting on the casting.

Castings may be hindered in their contraction by (1) resistance of the mold, and (2) the shape and form of the casting (see Chap. VIII). There is sufficient evidence to show that internal stress arises from the prevention of normal contraction by the rigidity of sand molds and cores. A simple example will suffice. Figure 165 represents a hard dry-sand mold of a bar with flanges on both ends. It is evident that the sand between the two flanges will prevent the cooling bar from contracting normally. The bar tends to contract toward its center, but the sand pressing against the inside faces of the flange prevents it from so doing, and hence stresses are built up as the casting cools with the bar portion under tension.

In all probability an external hot tear will form at the junction of the bar and flanges. The reason the casting tears at these points is that the
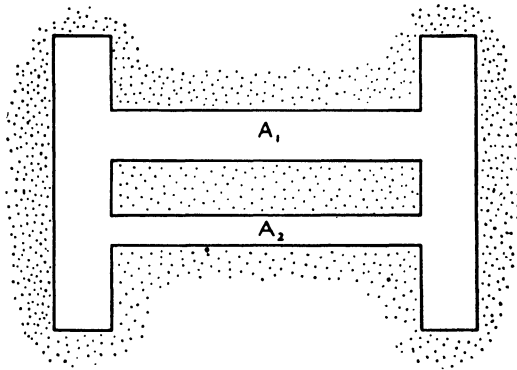
Fig. 166.—Two bars joined by flanges.

tension stresses developed in the bar by the flanges will concentrate at a hot spot, which is the weakest section and in this case is the junction of the flanges to the bar. The stresses developed in this case were sufficient to exceed the mechanical strength of the metal at that particular temperature, and hot tears were formed.

The location of hot tears is nearly always at points of design weakness, such as at reentrant angles, at the joining of sections, or in sections made weak by the presence of hot spots. Hot spots, of course, are sections of extra mass usually found at junctions of joining sections. Thus, if hot tearing occurs, it will be at the weakest section, which presumably will be the hottest section.

It has also been stated that hot tears result from the shape or design of the casting. This is illustrated by Fig. 166, showing a design of two bars with flanges, in which a large bar of cross section A_1 is joined to a small bar of cross section A_2 by heavy end sections. Bar A_2 will completely solidify before the large bar A_1; thus a temperature gradient is

produced within the casting. This temperature gradient creates stresses in the casting, since the amount of solid contraction depends on the temperature and the hindered contraction.[3] The internal stresses created by these temperature gradients are so great in this case that the elastic limit of the material is exceeded, an easy matter, since the elastic limit at a very high temperature is very low and deformation is principally plastic.

The great difference in temperature is caused by the mold-chilling effect and the variation in cross-sectional areas of the two bars. Since the contraction of the steel is closely associated with the temperature, the normal rate of contraction of the two bars will be different. A possible
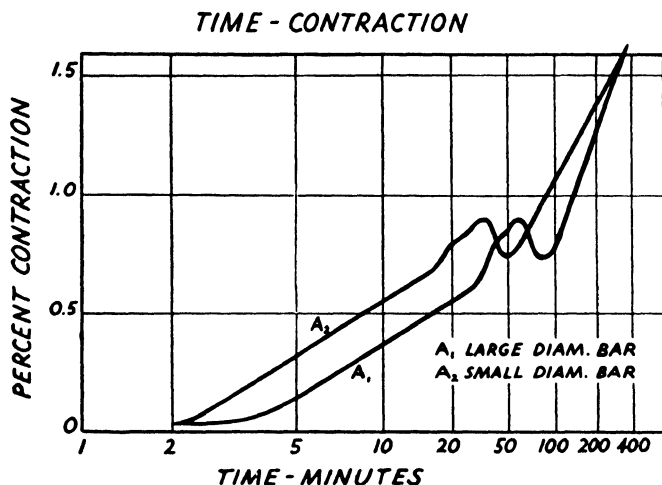
FIG. 167.—Time-contraction characteristics of bars of different sizes, not joined together. (*Briggs.*[15])

condition that may exist is shown in Fig. 167. The data plotted here represent the time-contraction characteristic of each bar, separated instead of joined by the flange construction. The greatest difference occurs shortly after the steel has solidified. Evidence exists[4] that there may be temperature gradients as large as 350°F. in a casting of intricate design.

Since the two bars are joined, they cannot contract, as shown in Fig. 167, but will take an intermediate course, with the small bar in tension and the large bar in compression and with the stresses mounting as the temperature drops. At some temperature the stress acting on the casting is greater than the strength of the steel, and the casting fails from hot tearing. The tears in this case will probably be found at the junction of the small bar and the flange, since the small bar is under tension and the junction is a weak spot and a position of concentration of stresses. Thus,
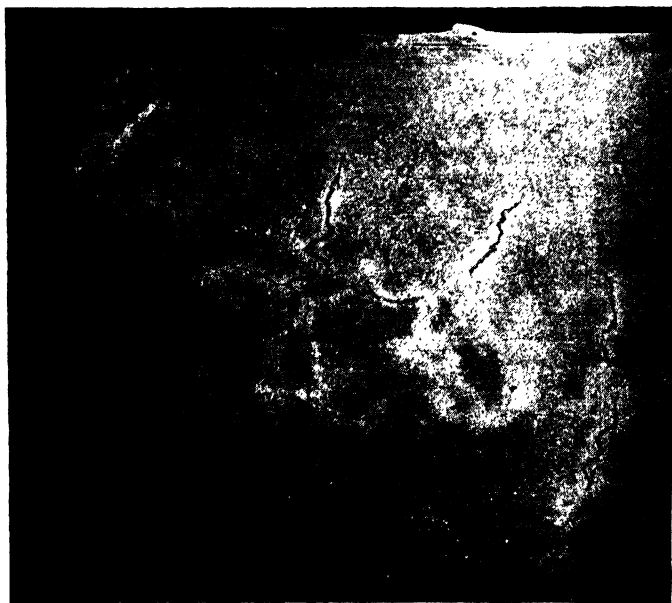
FIG. 168.—Internal hot tears brought to the surface by machining a steel casting. (*Courtesy of Pittsburgh Station, Bureau of Mines.*)

FIG. 169.—Radiograph of internal hot tear in a steel casting.

when steel is cast under conditions that tend toward high resistance to contraction, either because of mold resistance or casting design, the stresses acting on the casting are high, the rate at which they build up is fast, and the possibility of hot-tear formation is great.

Occurrence of Internal Hot Tears.—Internal hot tears are the result of two effects: (1) conditions similar to those causing external hot tears;
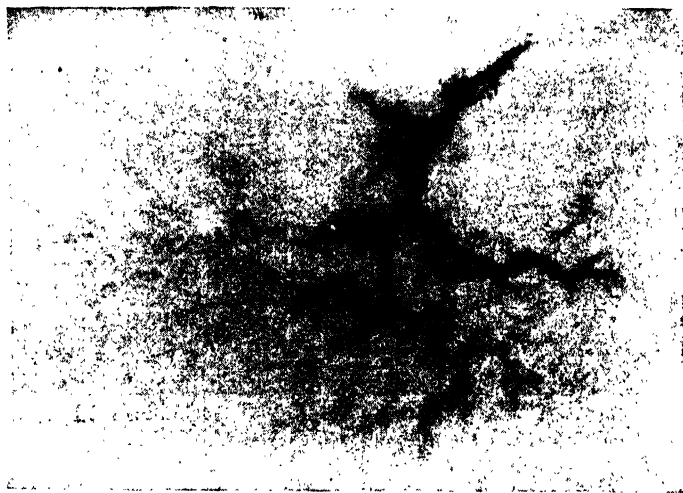
FIG. 170.—Radiograph of an internal hot tear resulting from solidification contraction. (*Courtesy of U.S. Naval Research Laboratory.*)
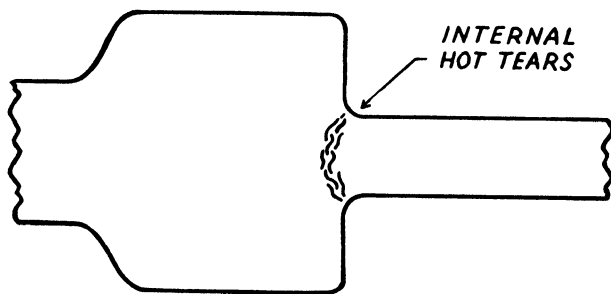
FIG. 171.—Formation of internal hot tears at a point of stress concentration.

viz., marked temperature gradients that cause solidified sections to pull away from partially solidified sections or completely solidified sections at a higher temperature (Figs. 168 and 169); and (2) solidification contraction which, because of insufficient feed metal, may result in the formation of hot tears rather than the usual shrinkage cavities (Fig. 170).

Internal hot tears are caused by stresses built up by temperature gradients. The same stresses cause external tears, but in this case the tears start near the center of the section and extend outward. The tears

do not come to the surface of the casting because the skin of the casting is at a lower temperature and is able to withstand the stresses. If it is recalled that one section of a casting may be contracting more rapidly than another, since the design of the casting is such that it is an enclosed stress-active system, then it is easy to see that tearing may take place in the weak central section of a casting at points of stress centralization without those cracks coming to the surface (Fig. 171).

It is also quite possible that this tearing may take place at temperatures higher than those found when external hot tearing occurs. In fact,
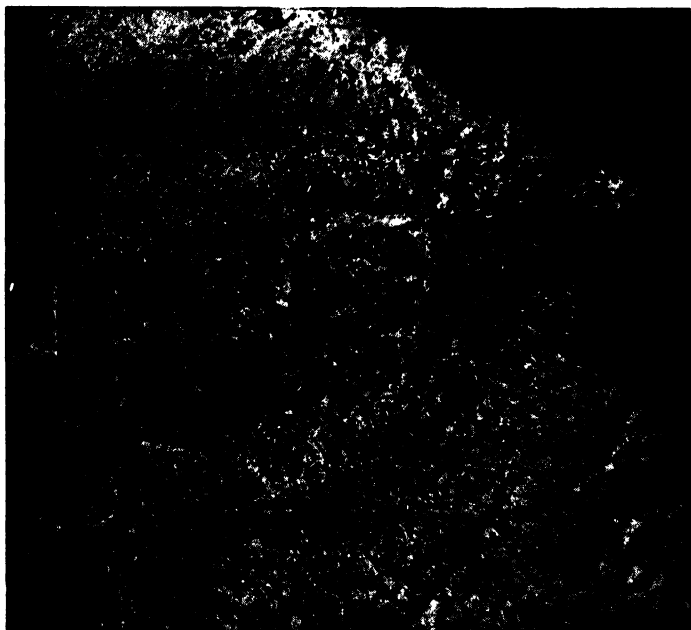
Fig. 172.—Hot tear subsequently filled with molten steel. (*Courtesy of U.S. Naval Research Laboratory.*)

it may occur during the solidification of the casting. At least, Singer and Bennek[5] believe so, for they have shown how internal hot tears in the solidified portion of the casting can be subsequently filled with fluid steel in the unsolidified central portion of the section. Such conditions have been found elsewhere, as shown in Fig. 172. The tears are of such shape and of such random distribution that it is evident they occurred while the steel possessed very low mechanical strength.[6]

It is also quite possible that mold resistance may play a part in the formation of the internal hot tears. This is a point that cannot easily be proved or disproved. However, because of the internal hot tears' close proximity to abruptly changing sections, it is believed that the

action of temperature gradients in the formation of stresses is the more important.

It has been suggested that solidification contraction may be responsible for hot-tear formation. Failure to allow for sufficient molten metal to provide for steel contraction in volume at the time of its solidification is the reason for the occurrence of these types of hot tear, as shown in Fig. 170. In uniform sections and in isolated sections normal shrinkage cavities may form as tears radiating out from small cavities, since the contraction of the metal toward the solidifying centers is sufficient to create a stress that tears the metal solidifying last. This stress need not be large, since the solidifying metal must be in any case very weak. Which of the two types of hot-tear formation is the more important is difficult to say.

Strength of Steels at Hot-tearing Temperatures.—It has been stated that hot tears form at high temperatures. Accordingly, the temperatures and the mechanical properties of cast steel at these temperatures are important to know.

Prior to 1928, much had been said concerning hot tears and how to prevent them, but little was known of the actual conditions of hot tearing, such as the cracking temperature or the magnitude of the stresses causing hot tearing. In 1928, Körber and Schitzkowski[7] produced hot tears in flanged bars by hindered contraction and decided that the critical temperature at which hot tears are most likely to form is about 2300 to 2400°F. They did not, however, measure the stresses on the bars at the time of hot tearing. In 1932 and 1934, Briggs and Gezelius[2,8] made a study of the stress acting on a bar during hindered contraction. These bars did not break under the hindered contraction, since they were so designed as to prevent points of stress centralization. Since the bars did not break, this work did not show the actual amount of stress necessary to cause hot tears at the hot-tearing temperature, but it did show the load-carrying ability of steel under hindered contraction.

In 1936 and 1938, Hall[9,10] of Woolwich Arsenal presented data on the strength and ductility of steel at temperatures near the melting point. In the region of 2300 to 2400°F., he showed that the ultimate stress necessary to cause failure in 1-in. bars averaged, for cast carbon steel, from 1,700 p.s.i. at 2370°F. to 2,500 p.s.i. at 2280°F. There is no indication from this work that the temperature range of 2300 to 2400°F. is more conducive to the formation of hot tears than any other temperature range. In fact, the temperature-ultimate-strength curves show nearly a straight-line relation, for there is practically a uniform increase in the ultimate strength as the temperature drops.

Figure 173 presents curves showing the strength of different carbon cast steels at elevated temperatures.[9] The curves indicate that the

increase of strength with the decrease of temperature is nearly a straight-line function, and that the lower the carbon content, the greater is the strength at any given temperature. There does not appear to be much difference between the values of the 0.25 and the 0.42 per cent carbon
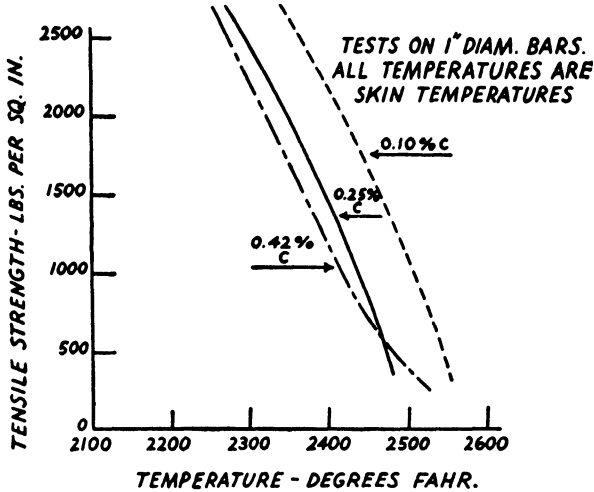
FIG. 173.—Strength of carbon steels at high temperatures. (*Hall.*[9])
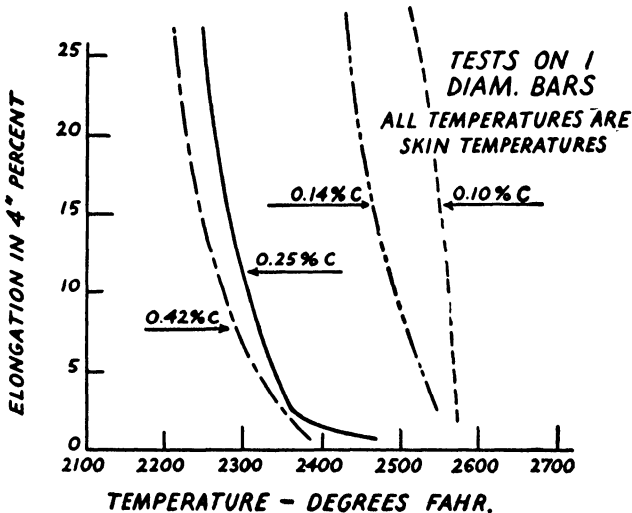
FIG. 174.—Ductility of carbon steels at high temperatures. (*Hall.*[9])

curves. This apparently indicates that as the carbon content increases, the strength differential diminishes rapidly. It would seem from these curves that the strength decreases as the temperature rises, thus making hot-tear formation easier.

The elongation values of the different carbon steels at the high temperatures are presented in Fig. 174. The curves show that as the carbon content increases, the ductility decreases for any particular temperature; the greatest difference in ductility occurs at the lower carbon content. An interesting point is presented by these curves; *viz.*, that for the carbon contents normally produced in commercial practice (0.20 to 0.40 per cent carbon) the ductility is very low until a temperature of approximately 2350°F. is reached in the cooling cycle.

Since there is little or no ductility in carbon steels until a temperature of 2350°F. is reached, then only a small amount of deformation is necessary for hot-tear formation. Castings that have failed because of hot
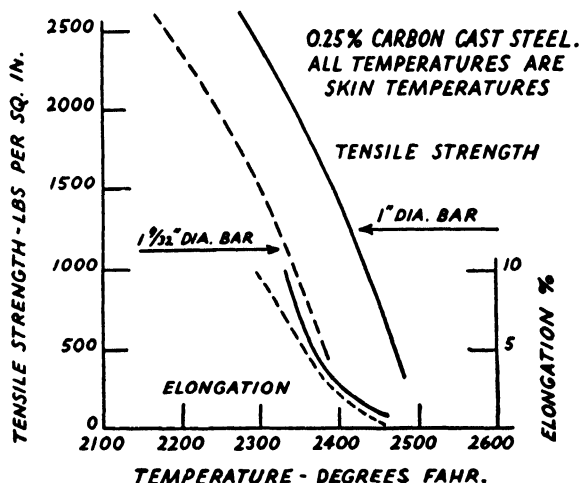
Fig. 175.—Effect of size of test bar on the strength and ductility of carbon steels at high temperatures.

tears exhibit sections the fractures of which show no evidence of plastic deformation. Ductility is, therefore, important in hot-tear occurrence. It may even be possible that it is more important than strength in the consideration of hot-tear formation. If this be so, then a temperature can be fixed below which hot-tear formation will probably not take place. For carbon contents of 0.25 to 0.40 per cent, temperatures of 2350 to 2300°F. would constitute the lower range of hot-tear formation. Hot tears therefore may form at any temperature between these temperatures and the solidification temperature of the steel.

Because of the increased ductility of low-carbon steels (carbon contents below 0.20 per cent), the range of hot-tear formation would be smaller by approximately 200°F., for ductility values have become appreciable at 2500 to 2550°F. This would allow an approximate temperature

range of only 175 to 225°F. for hot-tear formation for a 0.10 to 0.15 per cent carbon steel. For the 0.25 or 0.42 per cent carbon steel, a range of approximately 325°F. is available. This, of course, applies to 1-in.-diameter sections, and all temperatures recorded are skin temperatures.

In steel castings that solidify and cool in the normal way, a temperature gradient exists between the temperature of the skin and the temperature of the center of the casting. Thus, for example, in the case of a bar casting, the skin is at a lower temperature than the center of the bar, and since the lower temperature shows the higher property values, it must follow that the ratio of the area to the circumference of the bar is important in the magnitude of the values obtained upon testing bars of various sizes.

This is fairly well substantiated in Fig. 175, which shows the properties of 0.25 per cent carbon cast steel as tested by Hall[9] on a 1-in.-diameter bar and as obtained by the author on a $1^{19}\!/_{32}$-in.-diameter bar. The ratio on the two bars is as follows:

Diameter, in.	Area, sq. in.	Circumference, in.	Ratio
1	0.79	3.14	0.25
$1^{19}\!/_{32}$	2.00	5.00	0.40

From these figures it would appear that as the area increases with respect to the circumference, the effect of the skin on the properties of the bar is not so pronounced. The curves of Fig. 175 tend to prove that this condition actually exists.

The reason for discussing this point is that when mention is made of the magnitude of stresses at the hot-tearing temperatures, it should be remembered that it will not be necessary to develop stress forces equal to or greater than the tensile-strength values as set forth in the results presented by Hall. In fact, the strength of a 1-in. section in a casting whose width was greater than 1 in. would be less than the values shown by Hall, since the ratio of area to the circumference increases as the width increases. This condition is also true in section thicknesses that are greater than 1 in.

Magnitude of Stress Due to Hindered Contraction at Hot-tear-formation Temperatures.—It is known[2] that a 0.35 per cent carbon steel in the temperature range of hot tearing (2630 to 2300°F.) will develop stress loads under various types of hindered contraction (shown in Table XLVI) as follows:

TABLE XLVI.—STRESS QUALITIES AT HIGH TEMPERATURE

Stress, p.s.i.			Type of hindered contraction	Per cent contraction at room temperature
2500°F.	2400°F.	2300°F.		
105	.210	340	Light	1.7
335	500	625	Medium	1.0
385	560	730	High	0.5

These figures would tend to indicate that even under the most drastic conditions of hindered contraction the stresses developed during the hot-tear-formation temperature interval would not be sufficiently high to cause failure if the values of Hall are used as a basis of comparison. But it is known that castings do tear at these temperatures and, as has already been pointed out, sections in castings will not exhibit these high-strength values.

Tensile-strength values at elevated temperatures of carbon cast steel that had previously cooled at room temperature have been reported[11] to be 250 to 230 p.s.i. at 2250°F. These figures are more in line with the stresses developed in hindered contraction; but it is felt that not too much significance should be placed on these figures, since the temperature gradients usually found in steel castings were not present. The rate of testing used was much slower than that used by Hall and therefore more in accordance with the rate at which hindered-contraction stresses are built up.

Even though the tensile values that Hall shows are not practically applicable, the actual values of casting-section strengths at high temperature are undoubtedly still greater than the amount of stress developed by hindered contraction at the same temperature. This appears to require that stress concentration is responsible for the apparent rupture. It is quite possible that with the differential cooling of a casting, and because of hindered contraction due to mold resistance, stresses will concentrate at a hot spot in the casting with the result that the casting will tear. The hot spot may have a temperature approaching 2500°F., while other parts of the casting will be at temperatures of 2250 to 2300°F. In such a situation, stresses of 250 to 500 p.s.i. could arise through hindered contraction, which, if concentrated at the hot spot, would be—even according to the high values reported by Hall—a stress load sufficient to cause hot tearing. Thus, it is possible that small tension or shear stresses as developed by hindered contraction are responsible for hot-tear formation, providing that the hindered contraction present is greater than that normally allowed under "patternmaker's shrinkage."

Hindered-contraction Stresses Produced by Sands and Cores.—It is easily seen that casting design can be responsible for the stresses shown in Table XLVI, which is a brief summary of the results of Briggs and Gezelius.[2,8] But what is perhaps more difficult to understand is that sufficiently high stresses can be set up through hindered contraction from sand mold and cores to cause hot-tear formation. It can be demonstrated, however, that sands and cores can provide at elevated temperatures hindered contraction that is greater than that normally allowed under patternmaker's shrinkage.

Studies[15] conducted on the hot strength of molding sands and cores show that certain mixes develop very high strengths at elevated temperatures, as shown in Chap. XI. Some sands develop strength of over 1,000 p.s.i. in the neighborhood of 2000°F. It is true that they drop off to values of from 5 to 100 p.s.i. or so at temperatures near 2500°F. But this range of strength is found in a sand under the actual conditions of casting, since there is a pronounced temperature gradient in the sand following the casting of the steel.[16] Thus, while the sand at the mold-metal interface is at a temperature of about 2500°F. and has a strength of perhaps 20 p.s.i., the sand at a distance of only $\frac{1}{8}$ in. away will be at about 2000 to 2200°F. degrees and have a possible strength of 1000 p.s.i. Sand of these high strengths will, of course, tend to prevent the normal contraction of the casting, with the result that hindered-contraction stresses are formed.

From the work that has been done, it has been established that for steel castings the maximum hot strength at 2500°F. for sands probably should not be over 25 p.s.i. If this figure is exceeded by very much, the possibilities of hot tearing of the casting increase. Thus sands and cores, as well as casting design, can be responsible for the formation of high hindered-contraction stress at elevated temperatures.

Susceptibility of Various Cast Steels to Hot-tear Formation.—At any gathering of steel foundrymen where the subject of hot tears is discussed, sooner or later a debate will start as to which of the two types of cast steel—high carbon or low carbon—is less susceptible to hot tears. The discussion may also include acid vs. basic steel-melting practice as concerning susceptibility of hot-tear formation. Opinion among foundrymen is about equally divided on these subjects.

The odd thing is that during these discussions the subject of stresses or stress formation is seldom considered. Perhaps it is taken for granted, although it is believed that the subject is of such importance that it can scarcely be taken for granted, for without the presence of stress there will be no hot tears in well-fed sections, regardless of the composition of the steel or the manner in which it is made.

In the first place, the high- and low-carbon steels cannot be directly compared. Two similar castings produced under similar conditions will

not be under the same hindered-contraction stresses at a common temperature.   For example, assume that one casting is produced from 0.15 per cent carbon steel and another similar casting is poured from 0.45 per cent carbon steel.   At 2500°F. one can expect, under hindered-contraction conditions, a stress of 750 p.s.i. on the 0.15 per cent carbon steel whereas under the same conditions the 0.45 per cent carbon steel will have a stress of approximately only 500 lb. acting upon it.[2]

Because of the higher stress acting upon the low-carbon steel, the natural reaction would be that the low-carbon steel is more susceptible to hot tears.   However, this is not the entire story.   It must be remembered that the low-carbon steel has a much greater strength and ductility at 2300°F. than does the higher carbon steel.   If values established by Hall[9] are used (they can be considered comparative regardless of the section considered), the low-carbon steel has 1.4 times the strength and 8 times the ductility value of the high-carbon steel at 2300°F.   Also, the hot-tear-formation temperature range is considerably narrower than that of the higher carbon steel.   Thus the two types of steels cannot be compared directly as to susceptibility to hot tears.

In some cases, the high-carbon steels will be less susceptible to hot tears, while in other cases the low-carbon steels will resist hot-tear formation more effectively.   Which steel to use in order to obtain the best results will depend on the mold conditions and the casting design.   It must be remembered that hindered contraction can arise because of conditions of the mold, such as hard molds and cores, or it may reach considerable proportions because of the casting design alone.   It is suggested, however, that the low-carbon steels be used first if excessive hot tearing is encountered, since strength and ductility at high temperatures are necessary to combat stress concentrations at weak sections, which are at higher temperatures than other sections.

The question of basic vs. acid practice as to hot-tear-formation susceptibility is not so controversial.

Inclusions, their content, type, and characteristics present in a cast steel have considerable influence on the mechanical properties of the steel at the very high temperatures.   It is known that sulphur is an active inclusion former, and its presence under certain conditions can lead to the formation of harmful inclusion types.   Anything that could be done to keep the sulphur content as low as possible would be a step in the right direction in preventing hot-tear formation.   It has been shown[10] that as the sulphur content increases, the strength and ductility of cast steel at high temperatures fall off extensively (Fig. 176).   By the use of the basic practice it is possible to keep the sulphur content low.   Therefore, things other than sulphur content and inclusion count being equal, basic steels are less susceptible to hot-tear formation than acid steels.

Cast steels having a sulphur content greater than 0.03 per cent will show, upon deoxidation with a critical amount of aluminum (0.05 per cent), an extensive chainlike eutectic pattern of inclusions.[12] Such a condition results in low ductility. Sulphur content in the neighborhood of 0.02 per cent has but little effect in producing low ductility if the practice of deoxidation with small amounts of aluminum is used. Low ductility and low strength are obtained at high temperatures and are more pronounced than those recorded at atmospheric temperatures. Thus, from the standpoint of hot-tear susceptibility, the sulphur content should be kept low, preferably below 0.025 per cent.

CURVE B - CARBON 0.18 % SULPHUR 0.003 %
CURVE A - CARBON 0.14% SULPHUR 0.070 %
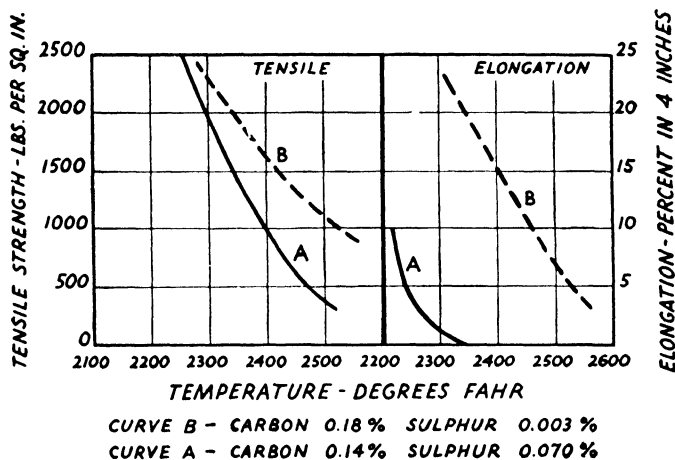
FIG. 176.—Effect of sulphur on the strength and ductility of steels at high temperatures. (*Hall.*[10])

High phosphorus (0.07 per cent) lowers the strength and ductility of cast steels at high temperatures, but normal phosphorus contents below 0.05 per cent do not have much effect on the high or atmospheric temperature properties of the steel.[9]

Inclusion Types.—Certain inclusions present in cast steel are conducive to the formation of hot tears.[3] Five types of inclusion have been recognized in cast steel in Table XLVII, together with their sensitivity toward hot-tear formation.[14]

The proneness to hot tearing of cast steels with silicate and eutectic inclusions is explained by the fact that atmospheric oxidation produces eutectic inclusions at the surface of the castings, which seriously lowers the resistance of steel castings to hot tearing. The intergranular-sulphide form of the ternary eutectic is considered to be more prone to hot tearing than the oxide-galaxy form. The presence of the galaxy eutectic adjacent to the oxide phase makes the galaxy type of steel less sensitive to hot tearing than the silicate or eutectic types.

TABLE XLVII.—TYPES OF INCLUSION AND SENSITIVITY TOWARD HOT-TEAR FORMATION
Crafts, Egan and Forgeng[14]

Type	Nonmetallic inclusions*	Hot-tear sensitivity
Silicate........	Glassy silicates, round sulphides (intergranular-sulphide films)	Prone to hot tear
Eutectic......	Round sulphides, intergranular-sulphide films (alumina)	Prone to hot tear
Galaxy........	Alumina, round sulphides, films of oxide galaxies (intergranular sulphides)	Less sensitive
Alumina......	Alumina, round sulphides (intergranular-sulphide films or oxide galaxies)	Resistant to hot tears
Peritectic.....	Alumina, duplex, round sulphides (intergranular-sulphide films)	Resistant to hot tears

* Parentheses indicate appearance only in small amounts.

The alumina and peritectic types are somewhat more resistant to hot tearing at the oxidized surface, owing to the formation of separated oxide particles rather than continuous intergranular films. Steels having eutectic inclusions (deoxidized with calcium) seem to have the same degree of resistance to hot tearing that has been found in the strongly deoxidized steels that are not treated with calcium.

These statements on inclusions should be looked upon as of minor importance, and there are other factors that require greater consideration. The use of a certain type of deoxidizer is not going to make the steel relatively resistant to hot tearing.

Influence of Pouring Temperature on Hot Tears.—The temperature of the steel as it enters the mold is not directly a responsible factor in hot-tear formation. Indirectly temperature may be a factor. If two steels are poured into similar molds, one a high-temperature steel and another a low-temperature steel, the high-temperature steel will not be more conducive to hot tears than the low-temperature steel. The reason for this is that hot tears do not form until the steel has solidified, and it is necessary for the hot steel to become cool steel before it solidifies. Therefore, there is no difference in the metal characteristics of the two steels as far as favoring hot-tear formation.

Temperature can, however, bring about altering conditions in the mold, and the casting design may be such that temperature of the steel contributes to hot-tear formation. It is known that large uniform-section castings should be poured on the cold side or as cold as the running of the mold will allow. In doing this the mold is not unduly heated at certain places to form hot spots in the casting. Points of change in sections are chilled to prevent hot-spot formations, points of concentration of stress. Castings of this type will probably develop hot tears if poured with high-temperature steel.

This does not mean that all castings poured with high-temperature steel are susceptible to hot-tear formation. In fact, by the use of high temperatures, it is possible to prevent hot-tear formation and at the same time prevent the formation of shrinkage cavities. If the casting is properly designed to allow for controlled directional solidification, then every possible attempt should be made to establish large temperature gradients within the mold and casting. One of the best ways to do this is to use high-temperature steel. Thus the effect that temperature has on hot tearing depends entirely upon the casting design and mold construction.

Alloy Steels.—In this discussion nothing has been said regarding alloy steels and how they react toward hot-tear formation. As in the case of carbon steels, the amount of strength and ductility that an alloy steel possesses at temperatures near the solidification temperature and the rate at which it contracts under hindered contraction will in all probability determine its position as to its ability to resist hot tears.

It is suggested that a comparison be made of a nickel-chromium cast steel of 0.35 per cent carbon content with a plain carbon cast steel of the same carbon content. It is universally held that a nickel-chromium steel is susceptible to hot-tear formation. From the information presented[9] it is possible to observe that nickel contents of from 2 to 5 per cent do not increase the tensile strength or ductility of carbon steel at high temperatures. Chromium in percentages of around 1 per cent appears definitely to lower the property values, while the combination of nickel and chromium reduces the values below those recorded for chromium.

If a few values are singled out for both carbon and nickel-chromium steels of the same approximate carbon content, the following is obtained, as presented in Table XLVIII:

TABLE XLVIII.—A COMPARISON OF CARBON AND NICKEL-CHROMIUM CAST-STEEL STRENGTHS AT HIGH TEMPERATURES

°F.	Carbon cast steel		Nickel-chromium cast steel	
	Tensile str., p.s.i.	Elongation in 4 in., per cent	Tensile str., p.s.i.	Elongation in 4 in., per cent
2500	400	0.0	60	0
2400	1300	0.5	380	0
2300	2000	7.0	840	0
2200	3200	33.0	1900	15

From these figures it can easily be seen that the nickel-chromium cast steels are much lower in strength and ductility than a corresponding plain carbon steel. Also, the hot-tear formation-temperature range is longer.

There appears to be but little difference in the rate of hindered contraction of carbon and the nickel-chromium cast steels,[8] and hence the rate at which the stress builds up is practically the same.   Under a hindered contraction that may be normally expected in casting manufacture, a stress of approximately 500 p.s.i. may be obtained at 2300°F.   This is very close to the 840 p.s.i. at zero elongation as shown for the properties of the nickel-chromium steel, as compared with the values of 2,000 p.s.i. and 7 per cent elongation for the carbon steel.   Hence it is easy to see that the nickel-chromium cast steels are very susceptible to hot-tear formation.

It has been stated previously that solidification contraction may be responsible for the formation of hot tears.   The information on the amount of solidification contraction in various steels is rather limited, and one is called on to make certain assumptions after reviewing all the assembled data on specific volume values.   Thus it appears that the total solidification contraction of a nickel-chromium steel would be perhaps slightly greater than that of the carbon steel.   If there has been no error in this deduction, then poorly fed nickel-chromium steel would be more susceptible to hot-tear formation (contraction cavities) because of insufficient feed metal than the plain carbon cast steel.   Looking at the problem from these two viewpoints, it is easy to see why the nickel-chromium combination is disliked in the foundry.

Methods of Testing for Hot-tear Susceptibility.—It would be possible to continue at some length on the subject of hot-tear-formation susceptibility, but a considerable amount of the information would be based upon general beliefs instead of actual information.   More important information could be obtained if steel foundrymen would carry out lines of correlated testing.   This means that some type of test must be developed for hot-tear susceptibility.

Several foundries in this country have attempted to develop a test bar but have soon realized that there appear to be many variables that must be carefully studied before the correct test bar can be devised.   Such items as the diameter of the bar and the length and shape of the bar must be considered carefully.   If the bar has a uniform section, it will not tear. The bar must not be too long or it will tear with all steel compositions. If it is too short, it may not tear at all.   Of course, the ends of the bar must be fitted with flanges, or something similar to flanges, so that the mold will hinder the contraction of the steel.

The essentials of the design should be a bar the length of which is probably between 1 and 2 ft.   It must contain a hot spot such as illustrated in Fig. 171 so that hot tearing will be possible.[10]   The diameter of the increased section will probably not be greater than twice that of the shank.

Measures for the Prevention of Hot Tears.—The best method of preventing hot tears is to keep hindered-contraction stresses to low values. If hindered-contraction values are greater than normal patternmaker's shrinkage ($\frac{3}{16}$ in. per ft.), hot-tear formation is a distinct possibility.

Measures that should be undertaken to prevent hot-tear formation are as follows:

1. Keep stresses caused by mold and core contraction low. Do this by using relieving blocks or friable material in the backing sand, or by the use of weak cores.

2. See that castings are so designed that they will not be closed, stress-active systems wherein temperature gradients may be responsible for large stresses.

3. Make ample allowances for solidification contraction so that every section may be well fed.

4. Produce high-quality steel of low inclusion content and of proper inclusion type.

5. Apply chills to proper changes in sections to prevent hot-spot formation.

6. Use ribs or cracking strips at places to assist in preventing stress concentration.

7. Keep the pouring temperature consistent with the steel-casting design.

The details of the above items have been considered in Chaps. VI and VIII.

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CHAPTER X

STEEL MOLDING SANDS AND CORES

SAND TYPES

Origin.—Sands as used in the production of molds and cores for the manufacture of steel castings may be formed in a number of ways, as follows:

1. Residual deposits, resulting from the weathering of rocks, such as sandstones, etc.

2. Glacial sands, produced by the grinding action of glaciers.

3. Stream deposits, consisting of water-borne sand and its deposition along stream banks or at places where the stream-carrying ability is reduced.

4. Lake deposits, consisting of sands carried by streams or glacial sands, which are settled in lakes where sorting and grinding of the sands proceeds because of action of the water.

5. Marine deposits, which are similar to lake deposits except that the milling action is often greater because of wave action.

6. Wind-blown deposits, caused by sand grains carried for considerable distances by the wind and heaped into dunes. Sand classification is governed by the wind.

The first two deposits listed are of primary origin, while the others are methods of accumulation and classification.

Sands produced by the weathering of sandstones may consist of rounded grains, and those produced from igneous or metamorphic rock would probably be angular. Sand grains produced by the grinding action of glaciers may be angular and of varying size, but this condition also holds for some river and beach sands if the grading and grinding action of the water has not been extensive. Sands found in marine, lake, and wind deposits may also be well rounded and classified according to size. Rounding of sand grains is undoubtedly a very slow process, since but few sand deposits show a large percentage of well-rounded grains. Of 58 river sands examined by Ries and Conant,[1] only 22 showed rounded grains, but few of these contained them in any abundance. Of 59 marine sands, only 19 showed rounded grains.

Shape of Grains.—Sand grains can be divided into single and compound grains. The former may be further subdivided into round, subangular, and angular.

Compound grains (Fig. 177a) consist of small grains that have become cemented by iron oxide, silica, or calcium carbonate after the deposition of the sand. These sands are not used in present-day molding for the production of steel castings, because they split up easily under high temperatures. They are usually fairly porous and absorb oil or water, neither of which adds to the bonding properties.

The rounding of single grains is caused by the abrading action of one grain by another. Well-rounded grains are not plentiful and do not frequently occur in the coarser sizes above No. 40 sieve. Also, the rounding rarely extends to the smaller sizes below No. 70 sieve. The best
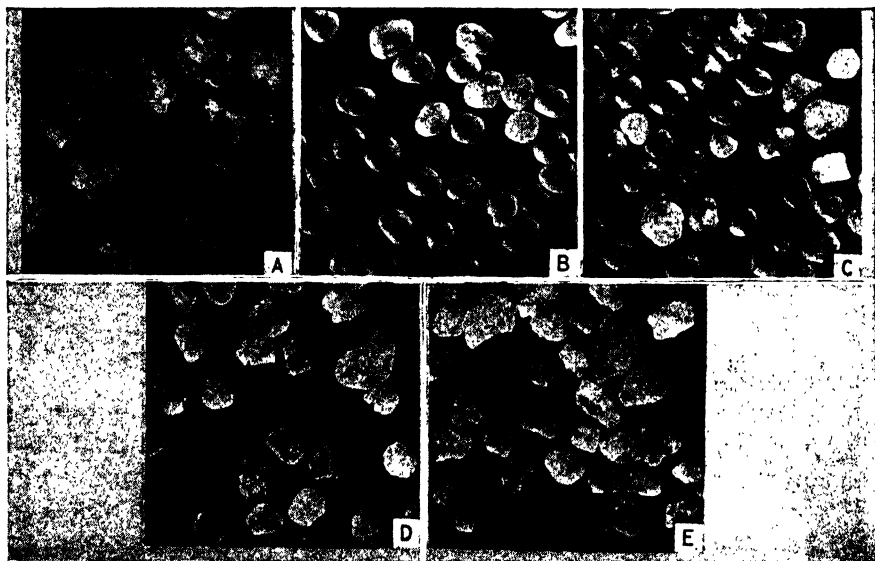
Fig. 177.—Sand Grains: (a) compounded, (b) well-rounded, (c) rounded, (d) semiangular, (e) angular. (*Ries and Conant.*[1])

known well-rounded sands are those of the Ottawa-Illinois district (Fig. 177b). The more generally found rounded types are similar to those of Fig. 177c.

Subangular grains (Fig. 177d) do not generally occur in any great quantity below the No. 140 sieve, but they can be found on all the coarser sieves. The angular grains (Fig. 177e) seem to predominate on the smaller sieves in all classes of sand, but often they are found in all sizes.

The surface of simple sand grains may be described as smooth, rough, glassy, frosted, pitted, and stained. All of these features may occur in the same sand, but they are not outstanding characteristics in the acceptance of the sand as a good molding sand. Sands having grains of a uniform size are likely to be more permeable than those with grains of mixed

sizes, if the grains are not too fine. With a mixture of sizes, the small grains pack in between the large ones and permeability is reduced. The shape of the grains affects the flowing qualities of a sand in that the more rounded grains are more flowable.

The mineral composition of the grains will affect the refractory qualities of the sand. Sands composed largely of quartz grains are highly refractory. Angular grains, with sharp edges and points, flux more readily than round grains, and the fine ones sinter more readily than the large ones.

The shape of the grain also exerts an influence on permeability and strength.[2] When bonded under the same conditions, an angular sand will show a higher permeability than a rounded sand, since the round grains pack together more tightly than do the angular grains. The rounded-grain sand is stronger[35] since there is a larger number of surfaces in contact, and thus the contact is better than can be obtained with angular sands. Also, the clay is more evenly distributed in rounded-grain sands, and there is always a film of clay between the grains. In angular sands, the contacts are very irregular and the points of contact may be extremely small. Angular sands should be mulled for long periods of time, in order that the bond may be better distributed over the grain surfaces. Hard ramming should be used to bring angular grains in close contact.

In the period of years around 1930, it was the general opinion of steel foundrymen that the well-rounded sands were the ideal molding and core sands. That idea has since changed, and by 1943 it was the accepted belief that the subangular-to-angular sands were the best molding sands. This change in thought followed the discovery that the angular sands made better mold-cavity surfaces in that they could be rammed harder, with the result that there were less cutting and scabbing. The rounded grain is still considered the ideal shape for most core sands.

The trend during the past few years has been to obtain a classification of sands, with the bulk of the grains caught on three adjacent screens with a small percentage on possibly two more adjacent screens. A sand of such a classification permits high permeability in relation to its fineness and also tends to give low expansion qualities.

It has occasionally been considered in the past, by various investigators, that all the fine sand particles—those of minus 150 mesh—should be removed from the sand. The present trend is to permit a small percentage of fines in the sand, to prevent metal penetration and buckling. The expansion of a closely packed, uniform-grain sand may be responsible for buckling. Fines reduce the flowability of a sand by increasing the distance between the sand grains. Also, the total expansion is not so great as when a small percentage of fines is used.

Types.—Many kinds and varieties of steel-molding sands have been used in the industry. These sands have been classified as natural-bonded sands, semibonded sands, synthetic-bonded sands, or mixtures of graded sands of any one of these. Some years ago it was not the exception to find in a shop as many as six steel-molding sands, which in turn were prepared into 15 different mixes.

Natural sands are those that are used by the foundry just as they are taken from the deposit. The clay content of these sands is from 5 to 20 per cent. Some of these sands are fairly well graded, with the clay seeping into the sand deposit from overlying clay deposits. The clay content of natural sands usually varies considerably; and with the advent of sand control to the steel foundry, these sands were dropped from use. Furthermore, the clay content is usually not very refractory, and these sands frequently contain a substantial percentage of silt, which adds little to the strength of the sand and reduces permeability. It is believed that not a single natural sand is now being used in the steel-casting industry.

Semibonded sands are those that contain from less than 1 per cent clay to about 5 per cent. They are frequently described as crude sands, since the sand producer seldom does much with them in the way of preparing them for foundry use. These sands are shipped to the foundry in much the same form as they were in the deposit. The sand technologist is able to control these sands as to their properties by the addition of clays, cereal binders, etc. Today many steel foundries use the semibonded sands because the small amount of natural clay present provides a priming surface on the grains and reduces the amount of clay to be added. Usually they are sands local to the foundry, and by building them up with the proper bonding materials they become acceptable to production molding and to production control.

Synthetic-bonded sands are those that consist of washed and graded sands, to which the foundry adds all the bonding materials necessary to obtain the desired molding qualities. A synthetic-bonded sand may be made from previously unused sand or from used molding and core sand. The addition of bonding materials to a semibonded sand produces a synthetic-bonded sand; thus all sands ready for use in the foundry are either natural- or synthetic-bonded sands.

In the rebonding of used foundry sands it is necessary to add unused sand to make up for that which was lost through breakdown to fines by heat and lost through adherence to the casting and removed in grinding or chipping reject material.

The final decision on the base sand to be used for the predominant proportion of the molding work will depend upon its delivered cost. The use of a semibonded sand may be preferable to a more costly washed and

  e. Oils
  f. Rubber and plastic compounds
 3. Inorganic
  a. Cements
  b. Silicates and esters

Clay-type Binders.—The clay substance of a molding sand is defined by the American Foundrymen's Association[4] as that part occurring in particles less than 0.02 mm. in diameter. The actual clay-flake particle is, however, of a size from several microns to less than 0.1 micron in diameter (1 micron is about 0.00004 in.). Nearly all clays are composed essentially of one or more of three minerals:[5] kaolinite, illite, and montmorillonite. In addition to the clay minerals, minor amounts of quartz, organic material, limonite, and other minerals are also found in many clays. Commonly known clays consist essentially of the following minerals: fire clays, usually mixtures of kaolinite and illite; kaolins and china clays, essentially kaolinite; weathered shales, usually composed of particles of illite; and bentonites, consisting of montmorillonite.

The properties of clays are related to the amount and nature of the exchangeable ions present. Hydrogen, sodium, potassium, calcium, and magnesium are the common exchangeable ions held by clays. Montmorillonite has about 10 times as much capacity as kaolinite to hold adsorbed ions. The capacity of illite varies, with some illite having the capacity of kaolinite and some several times this capacity. This power to adsorb ions is called "base-exchange capacity," and the adsorption capacity of the clay minerals can be determined by dye-adsorption tests.

The clay minerals differ from each other in their refractoriness and dehydration characteristics. Also, the montmorillonite clays have higher compression strengths than have the kaolinite clays. Some illite clays are weak, and others have high strength with the properties varying, depending on the exchange ions that the clay minerals contain.

Bentonite.—One of the most universally used binders in the steel foundry is bentonite. Bentonite[3] is a claylike substance formed by the weathering of volcanic ash and is composed of the minerals of the montmorillonite family. Two general types are known: the western bentonites, produced principally in the Black Hills region of Wyoming and South Dakota; and the southern bentonite, mined mostly in Mississippi. Both have about the same particle size and base-exchange capacity. They differ principally in the type of ion adsorbed on the surface of the clay particles. The western bentonite carries a high percentage of sodium ions and gives a slightly alkaline reaction whereas the southern bentonite has a high percentage of hydrogen and calcium ions and gives a slightly acid reaction.[6] The western bentonite can be distinguished from the

southern bentonite by the fact that it swells in water from 10 to 15 times its dry volume, while the southern bentonite swells similarly to other clays.

When mixed with sand, the bentonites differ greatly in the properties produced. Each has its own distinct advantages. The southern bentonite has not been used to any great extent in the steel foundry, although recent work by Dunbeck[6] and others has shown that it can be used advantageously since it has a low dry and hot strength, as compared with the western type (Table XLIX). It can be used in conjunction with western bentonite or alone for purposes requiring low hot strengths. Southern bentonite usually gives much higher flowability than the western bentonite.

TABLE XLIX.—DIFFERENCES IN WESTERN AND SOUTHERN BENTONITE
Dunbeck[6]

	Moisture, per cent	Green strength, p.s.i.	Dry strength, p.s.i.
5 per cent western bentonite.............	2.5	6.4	67.3
5 per cent southern bentonite.............	2.5	10.0	34.7

Since the western bentonite is the predominately used type of bentonite in the steel foundries, the author has followed the generally accepted practice of referring to the single term "bentonite" to represent the western type of bentonite. When southern bentonite is discussed, it will be referred to as "southern bentonite."

The bonding strength of bentonite in sand is from two to six times that of the natural-bonding clays found in sand or in the fire clays used for bonding sands. Bentonite has a fine particle size and stays in suspension readily. It also has a high plasticity and bonding power over a wide range of moisture content.

Fire Clays, Kaolin.—Fire clays and kaolin clays are used in synthetic-bonded sands as all or part of the clay-bonding substance. The minerals kaolinite and illite—found in fire clays and kaolin—are also found in natural-bonded and semibonded sands. Certain facts are known regarding the behavior of these clays. These facts may be enumerated as follows:

1. Properties of clays—such as plasticity, shrinkage, and bonding strength—vary with particle size. They tend to increase as particle size decreases.[7]

2. Clays with the same appearance and the same size distribution of composing particles will possess different physical properties such as bonding power, durability, etc.[7]

3. Clays having high base-exchange capacity also have high bonding strength. This relation is influenced by particle size.[7]

4. Low bonding strength is caused by the presence in amounts greater than about 30 per cent of such materials as quartz, calcite, and/or large clay-mineral flakes, which do not break down easily.[8]

5. There is a relation between the amount of clay mineral and the bonding strength in clays of the same clay-mineral content.[8]

6. The clay-bond distribution of a sand is a function of grain size as well as of milling time, kind, fineness, and moisture content of the clay.[9]

7. An increase in the silt content (materials other than clay but of a particle size smaller than 0.02 mm.) of a sand appears to lower its compressive and tensile strength. A fine-grained sand will stand the addition of more silt than a coarse-grained sand without serious loss of strength.[10]

The relative value of the various types of clay as to the properties produced is given in Table L. Comparison merely indicates the order from best to poorest in each property.

TABLE L.—RELATIVE VALUE OF VARIOUS TYPES OF CLAYS
Dunbeck[6]

	Kaolinite	Illite	Montmorillonite	
			Southern	Western
High green compression str	4	3	1	2
High dry compression str	2	3	4	1
High permeability	4	3	1	2
Least tempering water required	4	3	1	2
High sintering point	1	4	2	3
High durability	1	3	4	2
High flowability	3	2	1	4
High collapsibility	3	2	1	4
High resilience	2	3	4	1
Least contraction at 2500°F	1	4	2	3
Fewest lumps in shakeout sand	3	2	1	4
Least intensive mixing effort required	4	2	1	3

In natural-bonded sands there is considerable variation in the size-grade distribution within the clay substance of molding sands; in the relative amounts of quartz, clay minerals, and limonite that compose the clay substance of different sands; and in the size-grade distribution of these mineral constituents in the clay substance of various sands.[9] These variations explain to a considerable degree the absence of a close correlation between amount of clay substance and properties of natural molding sands.

It has been suggested that certain natural-bonded sands have greater durability than the synthetic sands bonded with montmorillonite.[10]

However, complete tests are necessary, since there are excellent and poor kaolin sands.

In the development of proper bonding, the practical manipulation of a foundry sand should not be overlooked. The amount and type of mulling given the sand, the method of addition of the bonding materials in the dry form or as slurries, and the amount of water addition can materially change the properties of the sand and the effectiveness of the bond addition. The influence of the items will be discussed in other sections of this chapter.

Organic Binders.—A number of organic binders are added to molding sands and cores. These binders may be added from a fraction of 1 per cent to as much as 2 per cent by weight. Nearly all organic binders increase the green strength of the sand and burn out at low temperatures.

Cereal Binders.—Cereal binders have been used for many years as a binder of sands and cores. There are several types: (1) dextrins, made from cornstarch; (2) gelatinized corn flour, made from dry milling of hominy grits or meal; (3) gelatinized starches, made from wet milling of cornstarch; (4) wheat flour; and (5) rye flour. In some cases a combination of a dextrin and a gelatinized corn flour is used.

A great many foundries use cereal binders in their sands and cores. The quantity used is in the neighborhood of 1 per cent by weight. These binders impart widely divergent properties to the sands, but in general they are added to prevent the sand from drying out too rapidly; or, if it does dry out, it will work better at a lower moisture content without crumbling than a sand with no cereal binder.

Some cereal binders increase the green bond strength, and most of them increase the dry strength. They also permit fast oven drying of sands. It has recently been found, through high-temperature-sand research, that cereal binders burn out quickly and thereby reduce the strength of sands at high temperatures. They also reduce high expansion values in certain sand mixes. They are used in cores, since only a minimum amount of core oil is required for the additional dry bond. The fast drying of cores containing cereal binders permits the reduction of drying cycles. Collapsibility of cores is increased by the burning out of the cereal binder.

Synthetic-bonded steel-molding sands are being used by some foundries without the addition of cereal binders. The properties of these sands have been reported in the technical press, and selected values are presented under Properties. Some foundries have found that certain of the binders are very useful, giving sand conditions similar to those found in the use of cereal binders as well as other additive features.

Lignin.—Lignin, a by-product of the sulphite liquor of the paper-pulp industry, has been used successfully by a number of steel foundries for

both sands and cores. It is obtainable as a powder or as a liquid, being water-soluble. Lignin has excellent binding properties with clays, is non-fermentative, and possesses rapid baking qualities at low temperatures. Lignins apparently do not impart much green strength to sands, but neither do some of the cereal binders. Lignins do have good resistance to high temperatures and permit rapid baking of sand. Lignins have the facility of producing hard mold or core surfaces without affecting the backing sand. In this use they may be advantageously substituted for molasses. They also have been used in cores to replace some of the core oil.

Lignin binders have found considerable use among those foundrymen who use semibonded sands and synthetic-bonded sands of fire-clay types. Tests of lignin binders show that it has excellent binding power with the clay minerals kaolinite and illite. The main disadvantage in the use of lignin binders is that they have a tendency to absorb moisture if molds or cores are permitted to stand after being made.

Resins and Gums.—The resins and gums have been used mainly as core binders, since they require heat to develop their best bonding properties. The gums are obtained by tapping trees and plants such as the yacca and distilling the extracted juices. The natural resins, such as pine resins, are obtained from shredded wood by steam and solvent extraction. Synthetic resins such as the petroleum and coal-tar resins are produced by chemical reactions. The synthetic resins are generally more costly than the natural resins and gums.

Resin-bonded baked sands resist moisture absorption and loss of strength in storage; they allow the core to break down rapidly; and low baking temperatures may be used. Some resins, such as the pine resin, are low in cost and can be used to bind molding sands with moderate strength.

A natural resin of the pine type is being used as a binder for molding sands, especially the dried sands. The use of the pine resin in quantities of about 1 per cent of the sand weight greatly increases the dry strength of molding sands over that obtained by the use of bentonite and cereal binders of the gelatinized corn type, or of bentonite alone. The resin also retards loss of strength in dry-sand molds exposed to a moist atmosphere.[36]

Pitch.—Pitch, a by-product of coke making, is used as a bond for cores, particularly large cores such as those used in pit molding. Such cores often require standing for long periods of time before being used, and the use of pitch as a binder is desirable since it is the most resistant to moisture absorption of any of the known bonds.

Protein.—Commercial proteins, gelatins, caseins, and glues are known as "protein binders," an example of which is soybean flour. These binders have good binding qualities, improve sand flowability, prevent

the sand from drying out too rapidly, and burn out at low temperatures. They are excellent when high collapsibility of cores is essential.

Oil.—The primary use of oil as a binder is in connection with cores. In most cases, steel foundries purchase prepared core oils that are produced with certain definite properties. The manufacturers seldom discuss the composition of their prepared oils. They may consist of drying or semidrying oils of vegetable, mineral, and/or marine animal. Any one oil or a combination of oils may be used. The oils may contain resin, metallic salts, dyes, etc.

Oil as a binder does not produce green strength, but it can give a wide range of dry baked strengths. At high temperatures oil burns out, allowing the sand core to collapse or break down readily. The properties resulting from core oils will be discussed under Cores.

Rubber and Plastic Compounds.—Small amounts of sand, such as those required for cores, have been bonded with rubber or with plastic compounds. Rubber latex is added to tempered sand, and cores are prepared shortly thereafter. The cores harden at room-temperature conditions. The use of other rubber compounds permits the low-temperature baking of cores. Plastics dissolved in alcohol are added to core sands. After the core is formed, the alcohol is ignited at low temperatures or the core is allowed to air-dry. Cores made with rubber or plastic compounds resist moisture absorption very well and have a quick collapsibility rate at steel-pouring temperatures.

Comparison of Organic Binders.—The dry binders have been compared by the addition of small amounts of binder to a standard AFA

TABLE LI.—GREEN COMPRESSION STRENGTH OF SAND AND ORGANIC-BINDER MIXES

	Green compression strength, p.s.i.				Moisture, per cent
	Per cent binder				
	0.5	1.0	1.5	2.0	
Western bentonite	0.69	0.99	1.40	1.78	2.66
Pitch	0.59	0.65	0.73	0.85	3.03
Resin	0.51	0.56	0.68	0.91	2.96
Gilsonite	0.55	0.57	0.59	0.60	3.00
Gelatinized corn flour	0.39	0.53	0.63	0.78	2.92
Glue	0.46	0.42	0.43	0.42	2.46
Lignin	0.42	0.40	0.40	0.40	3.00
Synthetic resin	....	0.45	0.44	0.46	2.97
Wheat flour	0.35	0.32	0.36	0.44	3.14
Rye flour	0.32	0.33	0.36	0.47	3.07
Dextrin	0.32	0.31	0.29	0.33	3.05

washed silica sand. Only sand, organic binder, and water were used in these mixes. A bentonite-bonded mix was used as a basis for comparison. Results on green compression strength and on toughness are given in Tables LI and LII. It is to be noted that the wholly soluble binders, such as dextrins and lignin binders, do not effect an increase in the green strength of a sand. Pitch and resin binders show the greatest green strength. Dry binders that give maximum toughness values are the gelatinized corn flour and the resin binders.

TABLE LII.—THE TOUGHNESS OF SAND AND ORGANIC-BINDER MIXES
Moisture, 3 per cent

| | Toughness (green strength × deformation × 1,000) | | | |
| | Per cent binder | | | |
	0.5	1.0	1.5	2.0
Western bentonite	31.1	47.5	64.2	78.2
Gelatinized corn flour	29.8	45.5	58.0	80.7
Resin	25.5	30.8	38.1	36.5
Rye flour	25.6	24.1	35.1	41.0
Wheat flour	21.0	32.0	38.5	35.2
Gilsonite	23.1	23.4	33.6	36.0
Pitch	23.0	26.0	27.0	34.0
Dextrin	18.4	22.2	29.3	33.0
Synthetic resin	....	18.9	20.6	20.4
Glue	16.6	16.4	17.1	21.5
Lignin	16.3	16.8	16.8	16.8

Inorganic Binders.—The inorganic-binder group consists mainly of Portland cement, used for molds and cores, and sodium silicate and silica esters, used with core sands.

Portland Cement.—For the past 45 years, Portland cement has been used as a binder in foundry molding and core sand. In recent years it has found rather wide acceptance in the steel foundry. The recommendations suggested by Menzel[13] regarding the use of Portland cement as a binder are presented under Molding Practice.

The advantages of cement as a bonding material have been listed[15] as

1. Lower material costs

2. Increased production under limited baking capacity, since cement-bonded cores and molds do not require oven drying

3. Elimination of flask equipment

4. Elimination of drying equipment

5. Mold surfaces are hard and abrasion and spalling of mold are at a minimum

6. Molds are not distorted through movement to drying furnaces, and the like.

Sodium Silicate, Silicon Esters.—Sodium silicate has been occasionally used as a binder for core sands and mold washes. It is a permanent bond in that it does not burn out at high temperatures. It makes a fairly dense bond and lowers the permeability of sands considerably. It does not permit the sand to collapse at high temperatures and lowers the fusion point of sand.

Silicon esters are being studied at the present time as potential bonding material for molds and cores. Their use is mainly in connection with experimental work on investment molding. Such studies have not been made public by the investigators. (See Molding Practice, page 358, for information on investment molding.)

ATMOSPHERIC-TEMPERATURE PROPERTIES OF MOLDING SAND

Sand Control.—Sand mixtures can affect the quality of the steel casting produced. The more closely these mixtures can be controlled, the more satisfactory are the resulting castings. It has become accepted practice in the steel casting industry to use only those sands and sand mixtures that will reproduce conditions and results for long periods of time. Sand control in the steel foundry is an absolute requirement.

By "sand control" is meant

1. Discovering by experimentation and observation the properties of sand most satisfactory for the castings being made and for the molding equipment available.

2. Making tests on foundry sands and core sands at sufficiently frequent intervals to keep informed on any changes that may occur in these properties.

3. Obtaining new sands and bonding materials, the properties of which are known and are uniform and whose effects on the foundry sands are understood.

4. Making the proper additions to the sand system as indicated by the tests, and properly incorporating these additions.

Careful control of foundry sands and a regulation of their properties have resulted in lower sand costs, fewer casting losses, and much improved casting surfaces. In order that sand control may be successfully applied in the foundry, all sands must be prepared by equipment that carefully measures, or weighs, the sand and the added bond and water. Milling times must be standardized and sand-handling equipment must be effectively controlled to prevent sand from drying out. Also, sand-reclaiming

systems must be provided and maintained, incident to the type of reused sand desired.

The operation of an effective sand control is such that trained personnel must be employed to see that the entire sand system functions properly. A sand technologist makes tests and maintains standards through changing compositions of sand mixes. Mill operators and reclaiming-equipment operators are necessary to a smoothly operating system. The three essentials of a sand-control operation are

1. Continual testing at specified intervals
2. Measurement of all ingredients
3. Timing of all operations

The principal routine testing in the steel foundry is the control of four variables: moisture, permeability, strength, and grain fineness. Other tests that are of assistance to the foundryman in controlling the sand quality are mold hardness, flowability, and deformation of the sand. Primarily, however, sand control consists of the standardization of grain size, bonding contents, and moisture. Tests have been prepared and standardized[4] for foundry sands and cores. Tests for molding sands are principally as follows:

1. Grain size
2. Moisture content
3. Permeability
4. Green strength
5. Dry strength
6. Flowability
7. Deformation
8. Specific gravity
9. Durability
10. Hot strength
11. Hot permeability
12. Contraction and expansion
13. Sintering point

Properties involving tests at temperatures above atmospheric will be considered in the next chapter. The control of these properties and the importance of such control will be presented in the following pages.

It is advisable that a few comments be made regarding testing and control. Tests on sands are not reproducible with sufficient accuracy to permit the practical use of specifications involving narrow tolerances. It is difficult to obtain uniform results by different testing laboratories.[17] This means that there are minor differences in individual procedures of testing. Full procedures of making every test, step by step, should be posted in a sand-testing laboratory; and checks should be made to see that sand technologists can check results of their own laboratories. All

materials should be carefully weighed and the moisture contents always recorded. Test specimens should be tested individually as soon as prepared, since surface drying affects results. A laboratory standard should be run regularly with other sands under test.

Moisture.—Water added to a clay-free sand will give to the sand a property of plasticity, permitting the sand to be shaped under various degrees of pressure, as is done by children in the building of sand castles on the seashore. The sand, however, is usually fairly wet, and when it dries, the molded object crumbles out of shape. When the sand is worked with insufficient or excess moisture, the casting surface is inferior in finish to that obtained when the sand is tempered to the best workable moisture contents. Moisture contents of molding sands are usually studied in conjunction with some other property of the sand.

Effect of Milling and Ramming.—The method of mixing the sand has much to do with the properties of the sand or with the amount of clay needed to bond a sand to a certain strength. The best conditions with the least amount of clay used are obtained when the sand mixture is intensively mulled. Such thorough mixing also results in sands of higher permeability, requiring less temper water, and of higher flowability than those sands mixed only and not mulled.

Fire-clay binders require the longest mulling; western bentonite mixes fast but has a tendency to form lumps if mulling is continued too long; southern bentonite also mixes fast, and there is no formation of lumps on continued mulling. By aerating sand after mulling, it is possible to reduce the density and increase permeability and flowability.[21] The correct mulling time for sands should be ascertained by test. Different types of mulling equipment will show different mulling times.

The amount of ramming that a bonded sand receives will affect the strength and permeability of the sand, as is illustrated in Fig. 178. As milling increases, up to a point the bond and moisture are more evenly distributed over the grains. Sands containing bentonite, cereal binder, or resin will have a lower permeability as the ramming is increased. A comparison has been made[36] between a 5 per cent bentonite, 5 per cent bentonite plus 1 per cent cereal binder, and 5 per cent bentonite plus 1 per cent pine resin on AFA 72 fineness silica sand with 4 per cent moisture as to the effect of ramming on permeability, strength, and hardness. The resin-bentonite bond showed the lowest permeability, with the rate of fall-off with increased ramming similar to the other binders. Green strength, green apparent density and mold hardness increase with the degree of ramming, with the bentonite–corn-flour mix the strongest. For a given number of rams the sand containing resin produced a higher strength, green apparent density, and mold hardness than that containing bentonite alone. Increasing the number of rams raised the dry strength

of the bentonite-bonded sand from 122 to 185 p.s.i.   The bentonite-resin mix was increased by additional ramming from 183 to 280 p.s.i.   Ramming increased the bentonite–corn-flour mix from 105 to 215 p.s.i.   Use
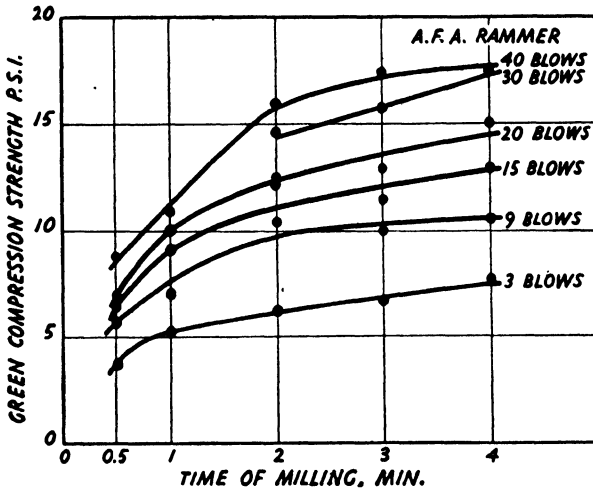
FIG· 178.—Relation between green compression strength and time of milling for different number of rammer blows.[22]
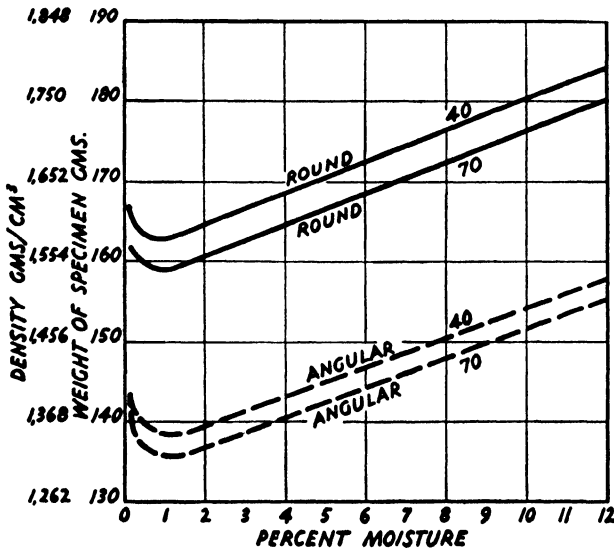
FIG. 179.—Effect of moisture on density of rammed green sand.   (*Ash and Lissell.*[18])

can be made of increased ramming and resin-binder additions to increase greatly the dry strength of dry-sand molds.

Density.—The curves of Fig. 179 show the relationship of moisture content with green apparent density of both round and angular sands.[18]

The density of a rammed-sand specimen is largely controlled by the frictional resistance between the grains. The degree of frictional resistance may be traced to the following factors: (1) size of the grains, (2) shape of the grains, (3) surface condition of the grains, (4) distribution of the grains, and (5) water content. The density increases with increasing grain size and moisture content. The rounded sand packs to a greater density.

The dry apparent density decreases as the intergranular space increases. As the water content is increased, it tries to surround the individual grains with a water film. The film grows thicker and stronger and increases the friction between grains until a critical film thickness is reached. Any increase in water content above that critical amount has no influence on friction. The critical amount of water is greater for a fine-grained sand because the surface area is greater. The dry apparent density of a sand is therefore constant and independent of water content
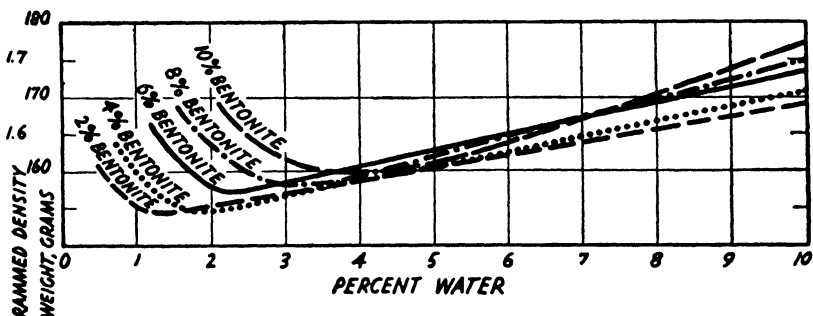
Fig. 180.—Effect of bentonite and moisture contents on the density of an AFA 63 sand.
(*Briggs and Morey.*[19])

above about 2 per cent moisture. The green apparent density, however, increases as a hyperbolic function.

The effect of bond (bentonite) and water contents on the density of sand is shown in Fig. 180. As the bentonite is increased, the weight to produce a rammed specimen exactly 2 in. high increases. By selecting the proper moisture content, a sand can be produced that will weigh less per unit volume than the same sand at higher or lower moisture contents. While this sand will have similar strength, it will contain more voids and will be more flexible as the metal cools and contracts. A poorly graded sand has greater weight than one with well-distributed grains. Well-graded sand (AFA 68), with 97 per cent falling on three screens, is from 6 to 10 per cent less dense than a poorly graded sand (AFA 83) with 83 per cent falling on six screens. Such a sand would have lower permeability and greater strength than the well-graded sand.

Permeability.—Permeability is that physical property that enables the molded mass of a sand to allow gas to pass through it. The permeability of a sand is influenced by

1. Actual amount of voids present
2. Surface area of the grains
3. Shape of the voids
4. Surface condition of the grains

These factors will be influenced by the sand-water-bond mixtures and the ramming method used, as follows:

1. Sand, according to grain size, shape, and surface contact
2. Water, according to the amount and surface tension
3. Bond, according to the amount and character
4. Ramming, according to the degree of compression of the sample

The relation between permeability and water content, for different green-sand sizes, is shown in Fig. 181. It will be noted that the angular
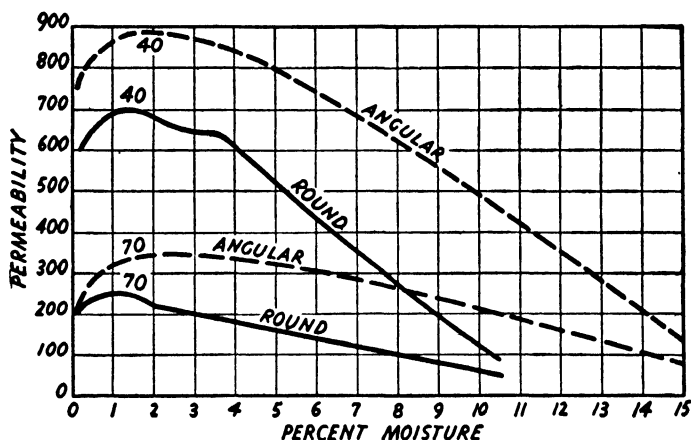
Fig. 181.—Effect of moisture on green permeability of round and angular washed sand. (*Ash and Lissel.*[18])

sand has a greater permeability and that it increases as the sand becomes coarser. Figure 182 compares the green permeabilities of a well-distributed sand and a poorly distributed sand. The permeability of the widely distributed sand (AFA 83) with 83 per cent falling on six screens is only about 60 per cent of that of the uniformly graded sand (AFA 68) with 97 per cent falling on three screens.

The effect of clay bond (bentonite) and moisture content on green permeability is illustrated in Fig. 183. It will be noted here that for each bond strength there is an optimum moisture content with respect to the permeability. Also, a sand-maximum distribution that falls on three consecutive screens will be less permeable than one that has a wide distribution on six or seven screens. The relation between permeability and AFA fineness number of steel-molding sands[23] is given in Fig. 184. If individual curves were plotted for each degree of fineness as a function of water content with a constant clay content, it would be found that all the

curves are similar in shape, increasing rapidly to a maximum and then decreasing approximately as a linear function of the water content. For bentonite-bonded sands, the maximum occurs between 2 and 3 per cent water, regardless of sand size. The magnitude of the maximum varies considerably.
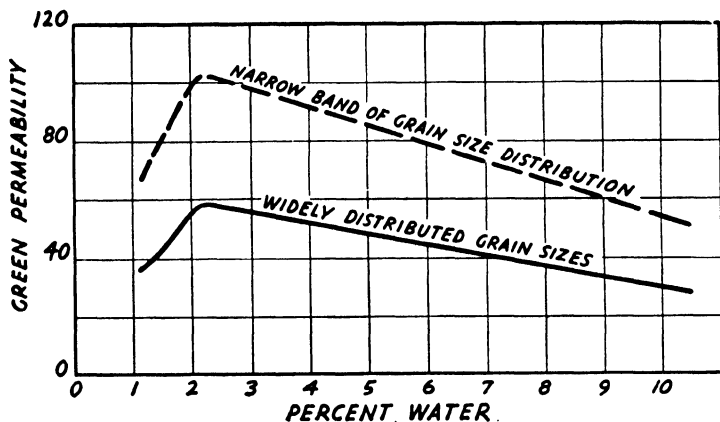
FIG. 182.—Effect of grain distribution on permeability in 5 per cent bonded bentonite sands. (*Morey and Taylor.*[20])
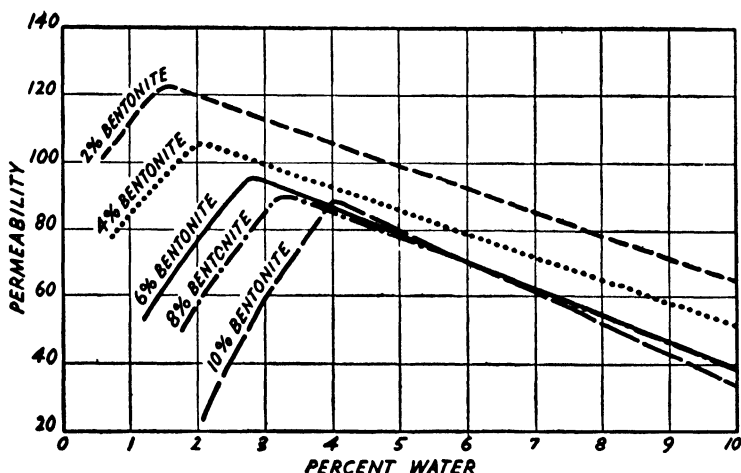
FIG. 183.—Effect of bentonite and moisture contents on the permeability of AFA 63 fineness washed silica sand. (*Briggs and Morey.*[19])

It is believed by steel foundrymen that steel-molding sands should have a permeability greater than about 75. This value will vary, depending on mold depth, mold hardness, moisture content, casting wall thickness, and weight of the casting. This value, or one greater, is advisable in view of the steam and mold gases that must escape from the mold. If,

however, the moisture content is maintained at a minimum and gas-forming binders are eliminated, the permeability could be much below this figure with excellent results.   It is general practice to have the backing sand of a higher permeability than the facing sand, in order to permit a natural flow of the gases toward the exterior of the mold.
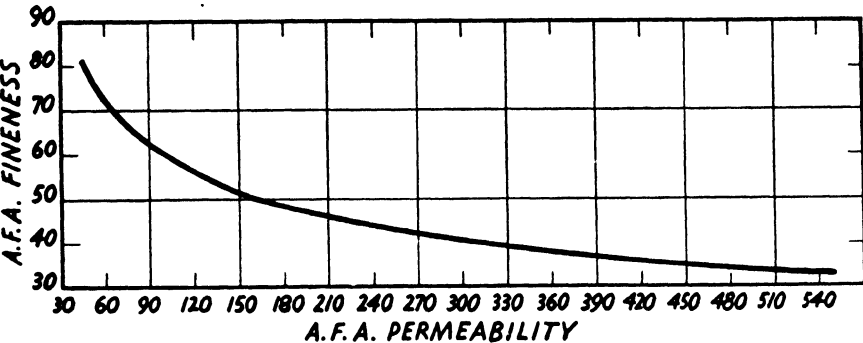
Fig. 184.—The relation between permeability and AFA Fineness Number.   (*Dietert, Woodliff, and Schuch.*[23])

$$\text{Permeability} = 0.37S \left(\frac{714}{M - 16} \right)^2 \text{ where } S = 1.0 \text{ for round sand grains}$$
$$M = \text{AFA fineness number}$$
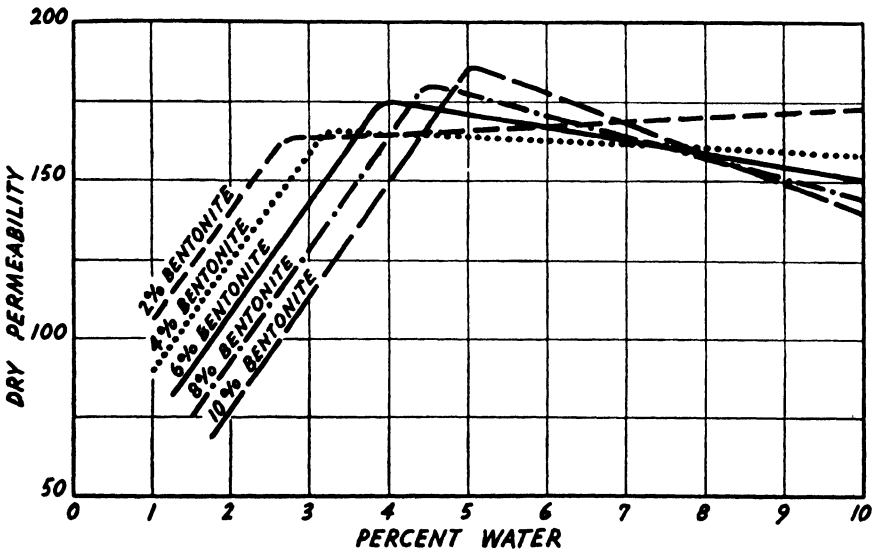
Fig. 185.—Effect of moisture in the green state on the dry permeability of AFA 63 sand-bentonite-water mixtures.   (*Briggs and Morey.*[19])

Dry-permeability curves are shown in Fig. 185.   In general, the permeability increases to a maximum with increasing water content and is nearly constant with further additions of water.   For bentonite-bonded

sands, about 1 per cent more water is required to produce the maximum dry permeability than to produce the maximum green permeability. Fire-clay- or illite-bonded sands present permeability curves similar in shape to those of the bentonite-bonded sands, except that the maximum permeability usually occurs at a higher moisture content, which is generally around 4 per cent. Mixtures of bentonite and fire clay will shift the maximum permeability to an intermediate value.

The addition of a cereal binder or a resin to a bentonite-bonded sand will result in a slight lowering of the permeability as the cereal binder or resin is increased to approximately 3 per cent. Further additions of
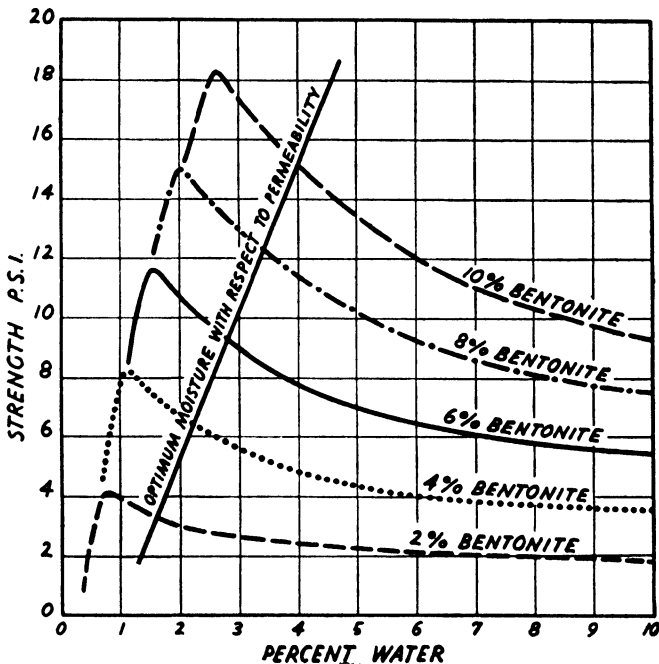
Fig. 186.—The relation between bentonite and moisture contents on the green compression strength of synthetic sand of AFA 63 fineness. (*Briggs and Morey*.[19])

cereal cause a remarkable increase in permeability[20] owing to tne voids produced between the pellets of cereal. With cereal binders of about 5 per cent, the permeability will increase rapidly and will be about twice that of the sand containing no cereal. Both green and dry permeabilities are similarly affected. Cereal binders, however, are seldom used in sand mixes beyond a 2 per cent addition.

Strength. *Compression.*—The compression strength of a sand is directly related to the clay content of the mix. The effect of varying the bentonite content of a washed silica sand (AFA 63) is shown in Fig. 186.

For each bentonite content, there is an optimum moisture content with respect to strength. This optimum value occurs at a moisture content that is below what is normally used in workable foundry sands. In the workable moisture range the strength falls rapidly at first as the moisture is increased. This graph may be used to design green-sand mixtures by drawing a horizontal line from the desired green-compressive-strength value, in the left margin, to the line labeled "optimum moisture with respect to premeability." From this point the bentonite content is found by interpolating between the nearest two bentonite curves, and the moisture is obtained by drawing a line vertical to the lower margin and reading the moisture from the scale. These values, of bentonite and moisture
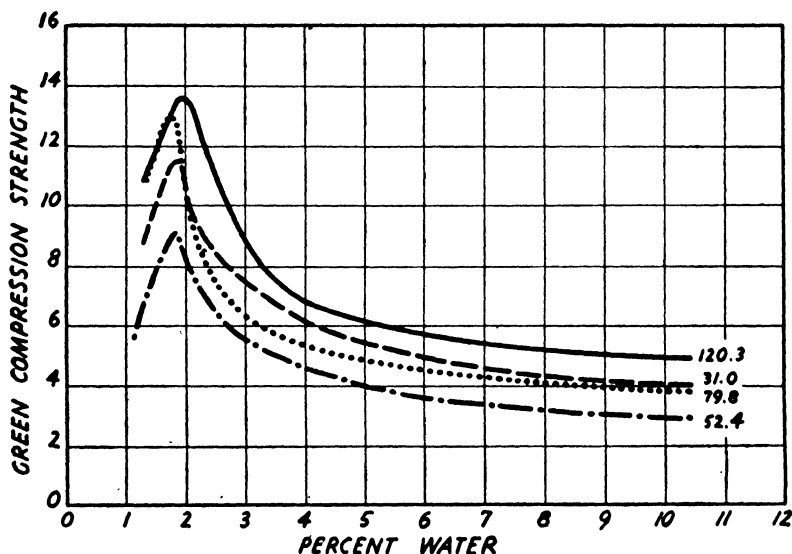
Fig. 187.—Effect of sand-grain size and water content on the green compression strength of washed silica sands bonded with 5 per cent bentonite. (*Morey and Taylor.*[20])

content, will produce the highest possible permeability at the desired strength. The daily addition of bentonite may be computed by plotting the green-strength and moisture content from daily tests on the graph. Interpolating between the bentonite curves gives the amount of effective bentonite in the mix; and subtracting this amount from the amount desired in the sand gives the amount of bentonite that should be added to the sand in order to maintain property control.

The effect of sand grain size on green-compression-strength values[20] is shown in Fig. 187. Sands illustrated were washed and well graded, rounded sands. In general, the highest strengths are obtained with the very coarse and very fine sands. This is because of two factors: (1) the surface area, and (2) the number of grain contacts. The coarse sands

have a relatively small surface area as compared with the fine sands, so that for a given percentage of clay the grains are covered with a thick layer, which tends to produce strength. The fine sands with a large sur-
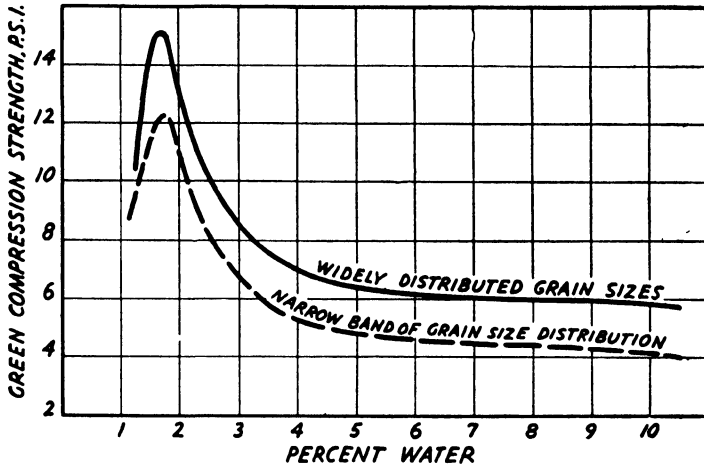
Fig. 188.—Effect of grain distribution on green compression strength of 5 per cent bentonite bonded sands. (*Morey and Taylor.*[20])
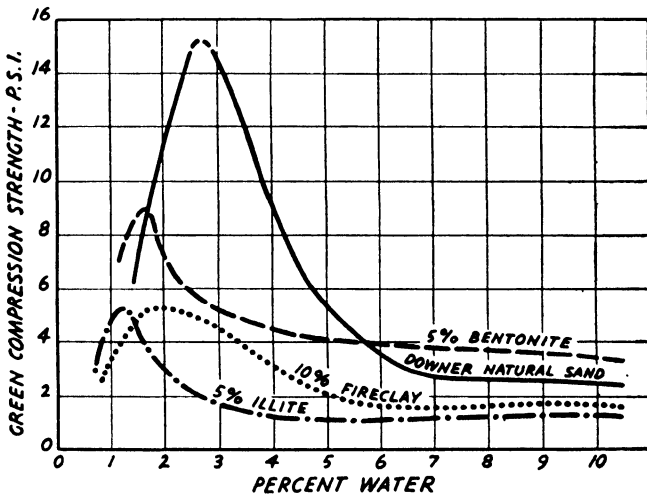
Fig. 189.—Green compression strengths of various clays as a function of the water content. (*Morey and Taylor.*[20])

face area have a thin layer of clay, but there is a much larger number of grain contacts, which contribute to the strength.

The size distribution of a sand affects the green compression strength,[20] as shown in Fig. 188. The sand with 85 per cent distributed over six screens is stronger than the 98 per cent graded to three adjoining screens.

Distribution has no effect on the shape of the curve or the position of the maximum strength, with respect to water content.

The effect of different clay contents on the green compression strength[20] is shown in Fig. 189. All clays are mixed with washed silica sand (AFA 63) with the exception of the natural-bonded sand. It should be brought out that in the fire-clay-bonded sands the moisture migrates freely from place to place within the mold. In bentonite sands, this migration is very
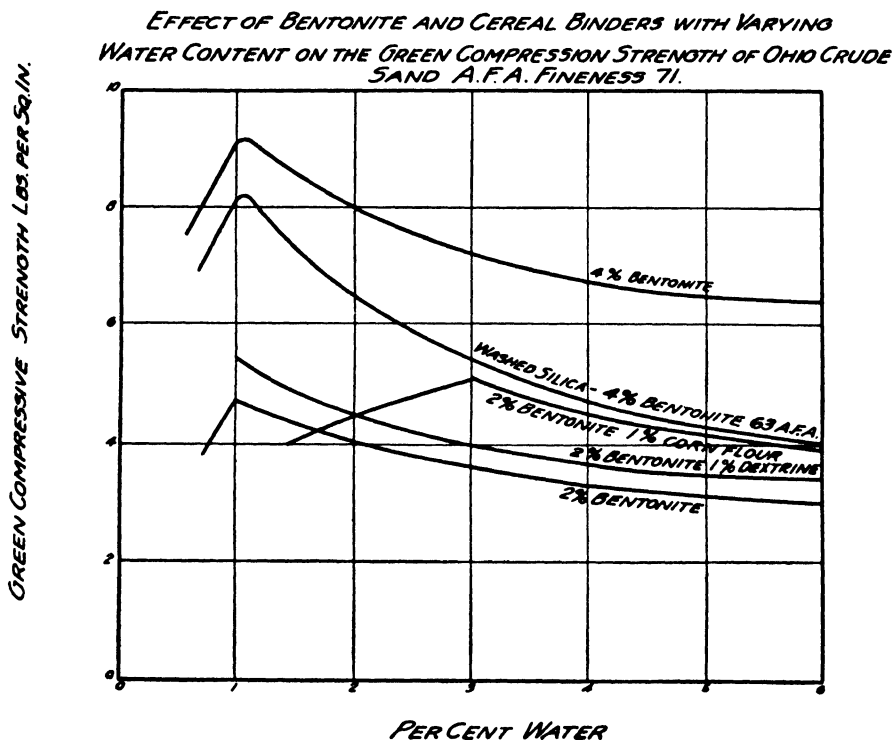
FIG. 190.—Effect of bentonite and cereal binders with varying water content on the green compression strength of Ohio crude sand of AFA 71 fineness. (*Caine.*[10])

limited; *i.e.*, if a completely saturated spot is next to a dry spot, the moisture distribution remains practically unchanged. This characteristic of bentonite leads to rapid surface drying, since the surface moisture is evaporated and is not replaced by moisture from the backing sand.

The green compression strength of clay-bonded sand increases as cereal binder is added to it. The strength of the mix may double in changing from zero per cent binder to 2.5 to 3.0 per cent binder with a constant water content. The increased strength is due not entirely to the strength of the cereal binder but partially to its effect on the amount of water available for the bentonite.

It is debatable whether a mix should contain sand and clay with cereal binder and a moderately high moisture, or the cereal should be omitted and less water used. Each method has its advantages. The omission of the cereal permits the reduction of water, which is always sought, but produces a dry and friable surface. The surface does not dry out so rapidly when cereal is used, and a harder face is produced. However, the use of a cereal binder results in introducing additional gases during metal pouring.
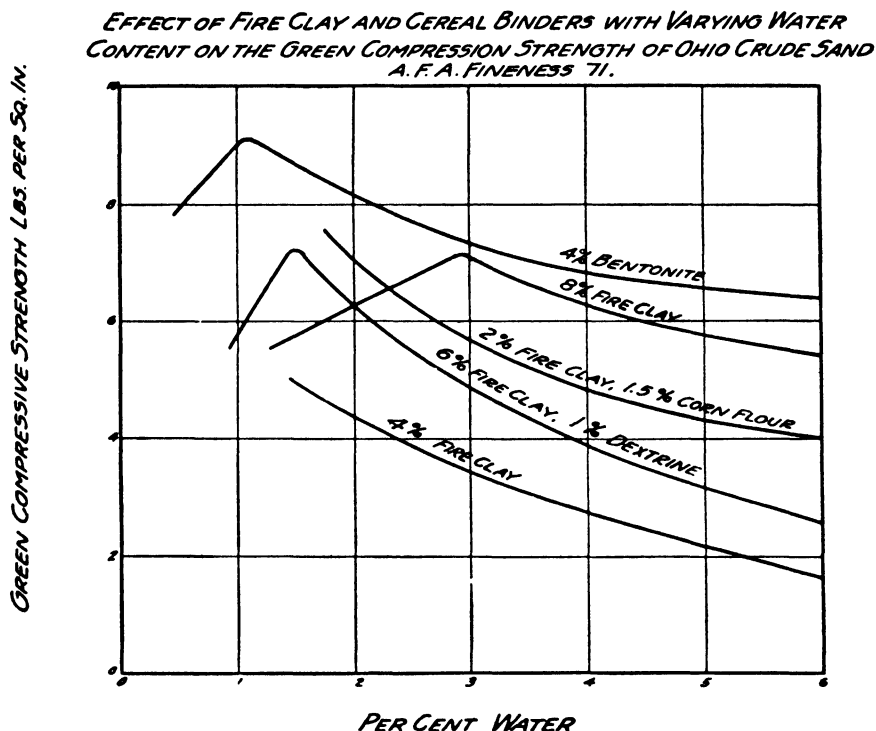
Fig. 191.—Effect of fire clay and cereal binders with varying water content on the green compression strength of Ohio crude sand of AFA 71 fineness. (*Caine.*[24])

Additional information on the effect of cereals on green compression strength[24] is given in Figs. 190 and 191. Gelatinized corn flour, when bonded with bentonite (Fig. 190), gives greater strength than a like percentage of dextrin. When cereal binder is added to fire-clay-bonded sand (Fig. 191), it is observed that the gelatinized corn flour increases the green strength appreciably, whereas dextrin does not. In order to keep the green strength constant, the fire-clay addition was lowered in the fire-clay–corn-flour sands. Instead of 6 per cent fire clay required with dextrin, only 2 per cent clay and 1.5 per cent gelatinized corn flour gave excellent green-compression results.

The relation between washed silica sand and a semibonded sand is shown in Fig. 198. The Ohio crude sand contains 3 per cent AFA bond substance. The great difference in green strength between the bonded crude and washed sand is due primarily to the natural colloids present in the crude sand and, secondarily, to the finer grain size.

The green compression strength does not seem to affect the performance of the sand when subjected to molten steel. If the sand has sufficient green strength to be moldable, that is probably sufficient. Increasing the green strength to values above this amount does not seem to improve the castings. It must be pointed out, however, that
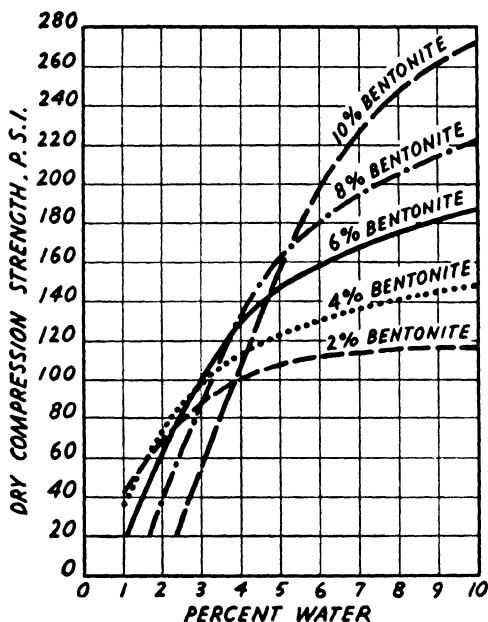
Fig. 192.—Effect of moisture in the green state on the dry compression strength of bentonite-bonded sands of AFA 63 fineness. (*Briggs and Morey.*[19])

a number of variables that increase the green strength also increase the dry strength and hardness of a sand. A study of sand conditions in steel foundries has shown that the green compression strength of sand varies from 2.5 to 11.5 p.s.i., depending on the type of molds desired. The strengths generally desired are from 4 to 6.5 p.s.i. The dry strength of sand may be increased by

1. More and longer intensive mixing
2. Increasing the moisture of the green sand
3. Harder ramming
4. Adding more clay and water
5. Adding various binders
6. Increasing the grain size and distribution

Oven-dried green sands produce strength properties differing considerably in value as well as in the shape of the curves produced. The relation between bentonite and water contents for a washed and graded silica sand is shown in Fig. 192. The dry strength increases rapidly as the moisture in the green state is increased. High bentonite contents produce very high strengths when the moisture content is high, but at low moisture contents they produce less dry strength than low-bentonite mixtures. Figure 193 is a combined graph showing green and dry com-
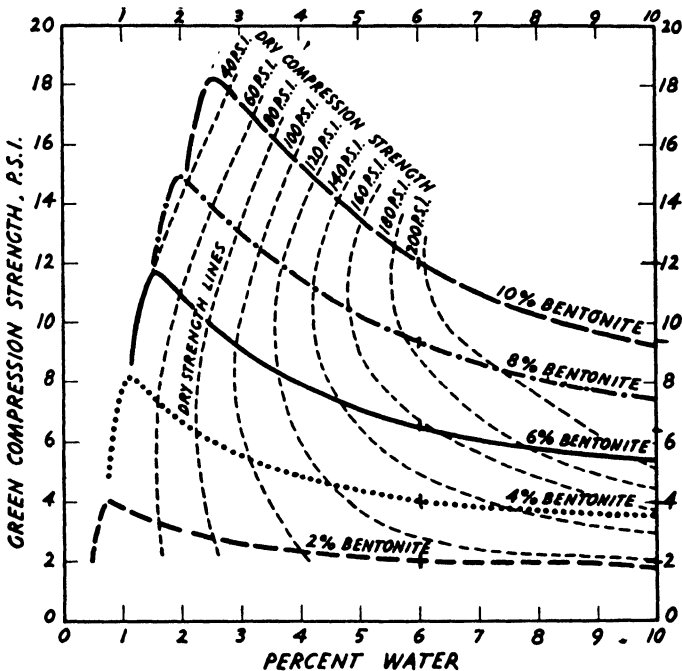
Fig. 193.— Green and dry compression strengths of a bentonite-bonded washed and graded sand of AFA 63 fineness. (*Briggs and Morey.*[19])

pression strengths for a particular washed silica sand (AFA 63) and shows the interrelation among four variables, *viz.*, water and bentonite contents, and green and dry compression strengths. When any two of these are known, the other two can be found from the graph.

A suitable dry sand can be produced from the synthetic green sand of the sand-bentonite-water type by increasing the moisture content when the sand is being mulled. This procedure does away with the necessity of making a special dry-sand mixture that must be handled separately in the shop. The finer sands produce the higher dry strengths. An AFA 120 sand is nearly twice as strong as an AFA 30 fineness sand, when bonded with the same bentonite content and at the higher moisture levels

(5 to 10 per cent water).   Also, a sand that is distributed over a number of screens will be much stronger than a well-graded sand distributed over a few screens at the same bond and water content.

The dry strength of fire-clay-bonded sands increases as the clay content increases, providing the moisture content is also increased.   They can be lower or higher than bentonite-bonded sands, depending on the amount of clay bond added.   The addition of cereal binders in quantities up to about 2 per cent increases the dry strength of a bentonite- or fire-clay-bonded sand considerably, especially with low water contents.   The dextrin binder is slightly more efficient than the gelatinized-corn-binder type in increasing the dry-strength properties.   The addition of pine-resin binder to a bentonite-bonded sand increases the dry compression strength of the sand considerably.[36]   Also, various sand compositions not containing resin will fall off considerably in dry compression strength if the dried molds are allowed to stand in moist atmosphere.   If pine resin is present in the mixture, in quantities of about 1 per cent, the falling off in strength is about half that of other mixes.

The use of southern bentonite in a mix will produce low dry strength. Cereal additions made to southern-bentonite-bonded sands will result in increasing the dry strength, the cereal addition overshadowing the type of bentonite used.

The hot strength (500 to 2500°F. testing temperature) of sand probably has more effect upon castings than the dry strength.   Dry strength is of some importance in the production of light castings when the surface-casting skin is formed before the development of any appreciable hot strength.[21]   Dry strength also has significance in relation to cracked castings and lumpy sand at shakeout, since much of the sand in most molds becomes hot enough for baking.

Tensile and Shear Strength.—The tensile and shear tests are not used so often as the compression test.   In fact, some foundrymen believe that these tests give no information that is not given by the green-compression test.   For brittle sands, however, the tensile strength is relatively lower than the compression strength.   It is the general belief that the tensile strength of fine sands (AFA 150 and higher) gives a more nearly accurate indication of active clay content than does the compression strength, since the fines in the sand create more compression than tensile strength.   For steel-foundry sands the green shear strength lies between values of 1 to 2.5 p.s.i., whereas the dry shear strength varies from 25 to 125 p.s.i. with an average of about 50 p.s.i.[23]

Hardness.—According to the American Foundrymen's Association,[4] "hardness" is the resistance offered by a sand mixture to deformation. The definition refers mostly to the degree of hardness of the mold.   The hardness of a mold is influenced according to the following:[25]

1. It increases with an increase in the strength of the sand.

2. It decreases with an increase in moisture beyond the well-tempered range.

3. It decreases slowly with an increase in fineness.

In bentonite-bonded sands of a constant bentonite content, but with increasing moisture contents, as the moisture content becomes greater, the mold hardness develops more slowly on air drying; but eventually a higher ultimate hardness is obtained. The addition of cereal binders to a bentonite mix will produce greater hardnesses and produce the hard-

THE EFFECT OF BINDERS WITH VARYING WATER CONTENTS
ON THE DRY HARDNESS OF OHIO CRUDE SAND A.F.A. FINENESS 71
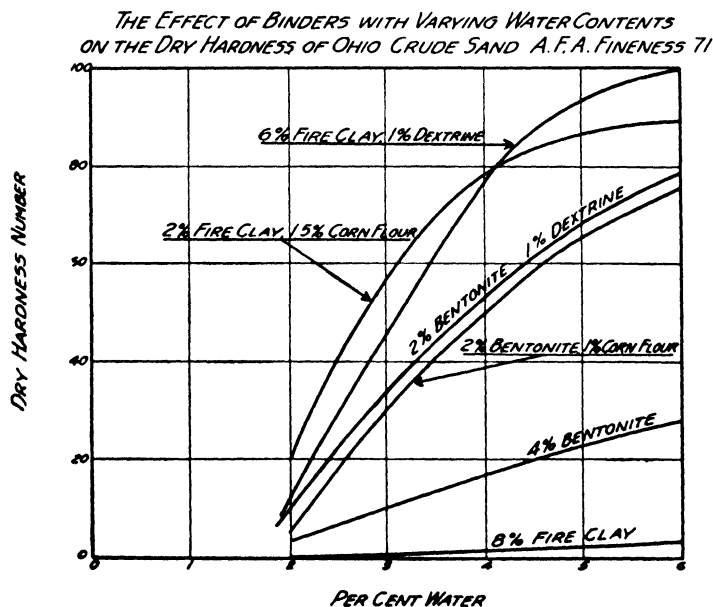
Fig. 194.—The effect of binders with varying water contents on the dry hardness of Ohio crude sand of AFA 71 fineness. (*Caine.*[24])

nesses sooner during the air drying of the mold than can be attained by the straight bentonite mix.[19]

No test has been devised by the American Foundrymen's Association that correlates mold hardness with the ability of the sand mixture to resist erosion or spalling by the action of the molten metal. A core-hardness tester, which employs the principle of scratch hardness, is available. Caine[24] employed a tester of this type on a crude Ohio sand bonded with various clays and binders, as shown in Fig. 194. It is Caine's observation that as the dry hardness increases, the amount of washed or eroded sand decreases. However, as the dry hardness increases, so also does the dry compression strength; and high dry compression strength may be responsible for cracking castings. With cast-

ings weighing over 100 lb., the dry hardness of the face of the mold cavity should be greater than 50 to prevent mold erosion.

It will be noted from Fig. 194 that cereal additions to clay and bentonite-bonded sand increase the dry hardness considerably and allow molding sands to air-dry firmly when immediate pouring after molding is impractical. The addition of 1 per cent pine resin to a bentonite-bonded sand increases the hardness of the sand.

Flowability.—Flowability is the property of sand that permits it to flow into the deeper pockets and corners of a pattern when a ramming energy is applied. The American Foundrymen's Association defines flowability[4] as "the mobility of a foundry sand by virtue of which it fills recesses and moves against pattern surfaces not necessarily in the line of squeezing or ramming."

The preponderance of the information as to sand qualities regarding the property of flowability has been obtained by use of the Dietert[27]
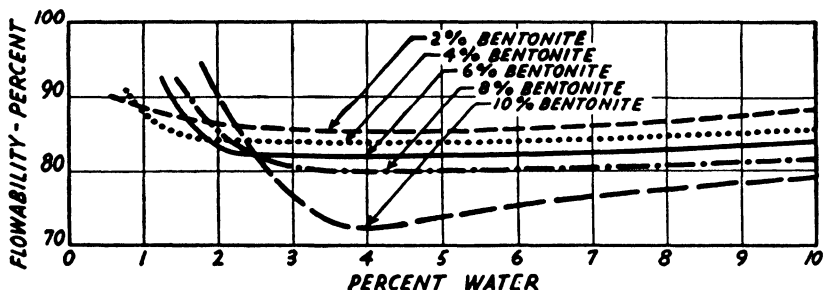
Fig. 195.—The relation of bentonite and moisture contents on the flowability of washed and graded sand of AFA 63 fineness. (*Briggs and Morey.*[19])

technique of measurement. The test established by Dietert is not without criticism,[28] and other investigators[29,30,31,32] have devised other tests. However, there seems to be general agreement as to the nature of flowability, and some disagreement as to whether or not all proposed tests measure this property. On the basis of data now available, it seems desirable to relate flowability to the preparation of a mold. Since steel foundries that measure this property are, for the most part, using the apparatus supplied by the Harry W. Dietert Company, a discussion of flowability will be based mainly on the use of this equipment.

A dry sand (without bond) has a relatively high granular friction and therefore flows less easily than a sand bonded with linseed oil or a diluted bentonite mixture where the binder acts as a lubricant. Water alone increases the friction between the grains and consequently increases the flowability.

The effect of bentonite and moisture contents on the flowability of a washed silica sand of AFA 63 fineness[19] is shown in Fig. 195. Similar results have been recorded by Lissell and Ash.[33] Flowability shows very

little variation in the normally used tempering ranges. There is a sharp increase in flowability as the moisture is reduced below 2 per cent. As the bentonite increases at a constant moisture content—for example, at 3 per cent moisture—the flowability decreases. In other words, as the green compression strength increases, the flowability decreases. Using the sand very dry increases the flowability, regardless of the amount of bentonite used.

As a sand becomes more uniform in grain size, it has a greater flowability when bonded with bentonite and water. Usually, however, a sand distributed over several screens with AFA fineness between 50 and 75 will have ample flowability for all practical purposes. Increasing the fire-clay content of a sand acts similarly to increasing the bentonite content in regard to flowability. Caine reports from his experiments that fire-clay-bonded sands have better flowability than bentonite-bonded sands. This does not always necessarily follow, and the reverse can be true when other sands are studied. By using an increasing quantity of gelatinized corn flour to a constant bentonite-water-sand mix, the flowability is decreased.

A study of steel-foundry sands[23] shows that the average flowability value is between 75 or 80 per cent, and that a steel-foundry sand should be maintained at a flowability of at least 75 per cent.

The relation between flowability and the other properties of sand is typical of many of the relations encountered in practice. Flowability is at or near the minimum value when other properties are such as to give the best molding practice. In bentonite-bonded sands, a moisture content of 3 to 4 per cent gives about the best combination of properties other than flowability. Higher flowability results in high dry strengths, which may cause cracking of the casting; also, scabbing and porosity may result. However, a sand with a minimum flowability will produce a good mold surface, and it is usually better to sacrifice flowability rather than some other properties. Uniform grain distribution is the best way of increasing the flowability if such increase is necessary. If southern bentonite were substituted for western bentonite, a further increase could be provided. Therefore, to increase flowability, (1) use sand of uniform distribution, (2) reduce fines, and (3) reduce the clay content.

Deformation.—Deformation is the amount that a sand "gives" or yields before it breaks. Deformation and flowability are certainly interdependent within any one mold. A poor flowing sand, which does not permit packing into a mold as hard as it should be, would cause excessive deformation under pressure from the molten metal. A sand of good flowability would give low mold-deformation values.

The amount of water and type of bond determine to a great degree the amount of deformation of a sand. The effect of bentonite and mois-

ture content with a washed silica sand of 63 fineness on the deformation value[19] is shown in Fig. 196. Deformation values are low for low moisture contents; but they increase as moisture increases. They are more pronounced when the bentonite content is high. As the grain size of a sand becomes smaller, the deformation values decrease. The addition of fire clay increases the deformation value, as compared with a similar content of bentonite.[20] Cereal binders added to a bentonite-bonded sand usually increase the deformation value, and this value increases rather rapidly as the cereal-binder content is increased. The addition of 1 per cent pine-resin binder to a mixture has little effect on the defor-
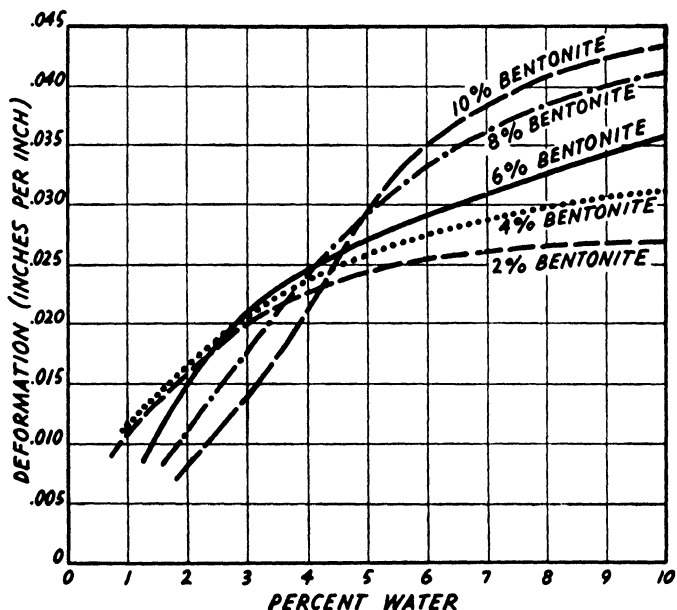
Fig. 196.—Effect of bentonite and moisture contents on the deformation of synthetic bonded washed and graded sand of AFA 63 fineness. (*Briggs and Morey.*[19])

mation property.[36] A uniformly graded sand has a higher deformation value than one that is distributed over a number of screens.[20]

The chart of Fig. 197 shows the working areas for various types of casting, as defined by deformation and green compression strength.[34] Workability of a sand should be controlled through deformation and strength to fall within the defined areas of Fig. 197. The curved line *BG* shows the limit to which clay-bonded sands can be worked. By soft ramming the workable range can be extended to include the area *B*. Beyond the green-sand-limit lines, skin drying or oven drying is required to stabilize the mold surface.

Deformation is not directly related to dry hardness, but generally it will be found to increase as the dry hardness increases. Deformation is

increased by less thorough mixing, increasing the moisture and clay content. High deformation may cause cuts and swells.

Toughness.—The toughness of a sand can be determined by multiplying the deformation values by the green compressive strength. A sand of low toughness is said to be brittle.

Since sand toughness is a product of two properties, it is affected by many elements within the sand mixture, and a variation of any one prop-
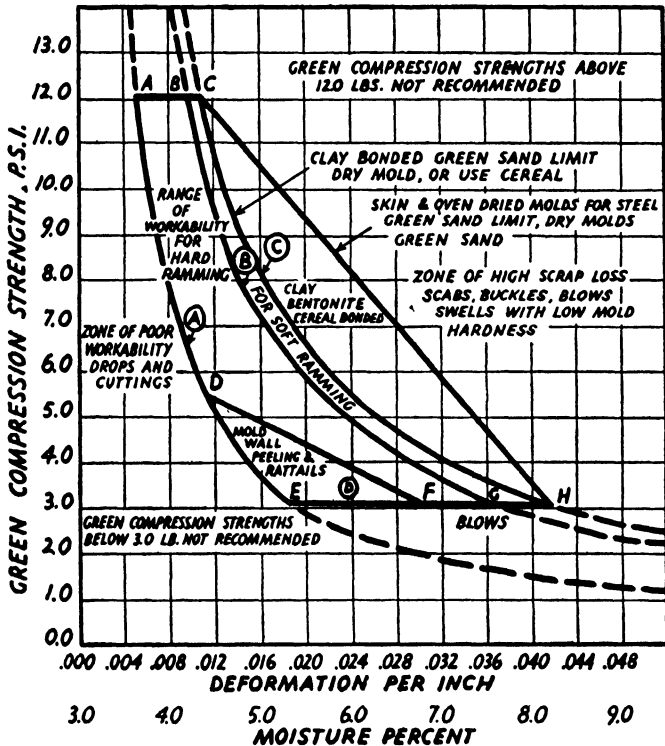
FIG. 197.—Deformation-strength-moisture diagram for foundry facing sands. For sands containing more than 10 per cent fines, add to moisture indicated in chart, fines 19.5. For high-plastic bonds and cereal binders or blends, divide moisture in chart by 1.8 for green steel-sand molding. For dry molding sand, add to this 1.5 per cent moisture. (*Dietert and Woodliff.*[34])

erty will enter into the mathematical relationship of calculating toughness value. Not too much work has been done in showing how variation in properties affects toughness. Usually this value is not calculated or used as a control value for steel-molding sands. A good working zone has been defined by Dietert and Woodliff.[34]

Value of Sand Properties at Room Temperature.—A study of the properties of sand at atmospheric temperatures is of importance primarily to assist in the making of the mold and in the selecting and con-

trolling of the sand mixture. The high-temperature properties of the sand mix are even more important in controlling and preventing the formation of casting defects from faulty sand conditioning. The present trend in the steel foundry is to give primary emphasis to the properties of sand mixtures at the elevated temperatures. Since the sands must be made and controlled as to uniformity at atmospheric temperatures, it is necessary that the properties so obtained at these conditions are understood even though in some cases there appears to be no correlation between atmospheric-temperature properties and elevated-temperature properties. Conditions of testing at elevated temperatures are not as yet thoroughly worked out, so that greater correlation than is at present at hand may be a possibility of the future.

It would also seem that it is not necessary or advisable to control all the properties of sand in a routine sand-control system. Which property or properties may be the most advantageous to control is not known for a certainty. The properties most universally watched by sand-control men are green compression strength and moisture content.

MOLDING PRACTICE

Type of Molding.—Steel castings are poured into a variety of molds such as green-sand, skin-dried-sand, dried-sand, cement-bonded, chamotte, and baked-core molds, etc. The choice of material is governed by such factors as the size and section of the castings to be made, the number required, design of the castings, facilities at hand in a particular foundry, economics of the production process, tolerances of molding, casting-surface characteristics, and probably other factors.

Green-sand Molds.—The green-sand method of molding is favored by foundrymen because

1. It is more economical than the dry-sand method of mold production.

2. It permits quantity production, since the molds may be rammed, closed, poured, and shaken out in a continuous cycle of short-time durations.

Practically all small miscellaneous castings are made in green-sand molds. The term "green sand" refers to the fact that no steps are taken to dry the mold after it is made. Green-sand molding is the preferred molding method for large-production runs of castings, even though the castings may weigh as much as 1,000 lb. Castings weighing 1 to 3 tons have been made in green sand and poured without subsequent heating. However, these molds often require many hours for preparation and assembling, and during this time the mold air-dries considerably.

Any method of drying molds, such as air drying, torch drying, or oven drying, adds to the cost of the mold preparation or to production delay,

or both. Since the economy of green-sand molding is an important factor in the production of steel casting, the progressive foundryman will analyze each design he is called upon to produce, with the thought of making it in a green-sand mold if possible.

In order to control and to minimize the effect of an excess amount of steam and gas, it is necessary that the most favorable conditions prevail with respect to sand practices. The sand control must be well organized, so that a uniformly bonded sand of good permeability, with a minimum moisture content permissible, is always available. It is essential that the heap, or backing, sand is controlled. Moisture content, strength, and permeability of the backing sand should be checked regularly. The permeability of the backing sand should not be lower than that of the facing sand, thus permitting an unhindered flow of gases from the mold.

The development of green-sand facing mixtures requires consideration of (1) metal-pouring temperature, and (2) method of ramming. Pouring temperatures of metal entering green-sand molds will be as high as 3000°F.; hence consideration must be given to the high-temperature properties of the sand. Facing sands for use on molding machines should have a flowability number of at least 80. The ramming of green-sand molds should be uniform and hard. The mold hardness should be from 85 to 95, as measured by the Dietert hardness tester.

Protection of the green-sand mold from undue metal erosion is afforded by the use of tile or core gates and hard-baked core inserts near the gate at the mold-cavity surface. By permitting green-sand molds to air-dry, the mold hardness is increased and the molds become more resistant to metal erosion.

Skin-dried Molds.—Green-sand molds that require additive protection from the erosion of the mold metal are dried at the mold-cavity surface. In order to produce good casting surfaces, especially in the thick-sectioned castings, it is often necessary that a more stable and a harder surface be prepared to withstand metal action. The surface of the mold cavity is dried by one of several methods, the most common of which is a gas torch. Hot-air drying, by use of portable stoves placed over mold openings, is frequently used.

A recent development is the use of infrared lamps to skin-dry sand molds. The adaptation of infrared lamps has been as an economy measure and space saver, where the quantity of orders did not justify installation of extensive drying equipment. The lamps are constructed in banks and made portable so that they can be shifted from mold to mold. They skin-dry the molds quickly and uniformly. Units of 150 lamps can be constructed to handle molds producing castings of 3 or 4 tons. The time required to skin-dry a mold 2 by 1 ft., with 24 lamps, to a depth of 1½

in. is about 90 min.   The method is suitable for skin drying provided the molds are free from deep cavities.

The mold-cavity surface is often washed with a refractory material prior to the skin drying to prevent metal penetration into the mold.   The sand is dried to a depth of about $\frac{1}{2}$ to 1 in.   The mold is closed shortly after skin drying and poured soon thereafter.   If skin-dried molds are allowed to stand for several hours after skin drying, the moisture in the backing sand will penetrate the skin-dried zone to produce a wetted area again.

Dry-sand Molds.—Dried molds are generally stronger than green molds and will therefore stand more handling.   They are usually used for larger and heavier castings than those molded in green sand.   These molds have a hard surface and evolve little gas or water vapor during casting.   With natural-bonded molding sands, the dry-sand practice is normally used, since these are not particularly suitable for green-sand molding.   In this country, synthetic sands are in almost universal use for the dry-sand practice.   They consist of fairly coarse-grained silica sands of the semibonded type, to which are added mixtures of bentonite and fire clay.

The molds are dried in large car-type ovens at temperatures up to about 500°F., depending on the mold mixture.   The ovens are, for the most part, oil or gas fired.   Drying time varies from 12 to 60 hr., according to the size of the mold and type of the oven.   The mold-drying temperatures vary somewhat with the sand mixture.   This can best be illustrated by a study of Table LIII.

TABLE LIII.—EFFECT OF MOLD-DRYING TEMPERATURE ON DRY COMPRESSION STRENGTH OF DRIED SYNTHETIC MOLDING SAND

Pragoff and Albus[36]

Materials by weight	Dry compression strength, p.s.i.		
	No. 1	No. 2	No. 3
Silica sand	95	95	95
Bentonite	5	5	5
Cereal binder	.....	1	.....
Resin binder	.....	.....	1
Moisture, per cent	4.1	4.0	4.0
Dried 2 hr. at 230°F	132	113	170
Dried 2 hr. at 400°F	122	104	209
Dried 2 hr. at 500°F	138	101	223
Dried 2 hr. at 600°F	129	81	200

The addition of resin increases the strength of dried-sand molds and likewise increases the drying temperature.   Ovens are more

easily controlled at temperatures of 400 to 500°F. than at lower temperatures.

The greatest efficiency in core and mold drying has been obtained by use of an oil- or gas-fired air heater. The air-heater units, mounted on top of the oven, develop a constant source of heated air at temperatures of about 500°F. The air and products of combustion are constantly fanned into the oven and returned in a major percentage of their volume to the heater for reheating and revitalizing with oxygen. Temperature gradients in this type of oven are not excessive, and there is plenty of air available to oxidize core oils.

Compo.— Molder's composition, more commonly known as "compo" or "Sheffield compo," is a synthetic molding mixture used in Great Britain.[37] Compo is probably the oldest of the synthetic molding mixtures. For many years it consisted of old crucible clay pots, old firebrick, and Sheffield ganister, all milled together with fire clay. In the days of crucible steel and crucible iron melting, the used crucible pots and firebrick were plentiful and cheap. In recent times these materials have become scarce. The material that has lately been adopted in place of the clay pots is a calcined bauxitic clay to which the name of "chamotte," from a similar material used in Continental Europe, is becoming attached. The compo mixture is used for large castings, upward to 120 tons in weight. Molds made of compo are oven dried before pouring.

Chamotte.—In Continental Europe a material somewhat similar to the English compo is used for dry-sand molding. The base is a calcined aluminous clay grog called "chamotte" which is graded to a size classification and milled with 10 to 15 per cent raw clay to form a molding mixture. A similar mixture has also been adopted by a few American foundries and is referred to as the "Fischer process" of molding.

The most important feature in preparing chamotte for molding is that the correct percentage of water be carefully added. If water additions of more than 4 to 8 per cent—depending on the clay content—are not made, the mixture may turn out to be a sticky mass that will be extremely difficult to mix and almost impossible to mold. Mulling times may be varied, but good results can be obtained in a period of 4 min.

The permeability of chamotte is very high, 2,000 or greater in the green state. The feature of high permeability is considered by European foundrymen to be essential in securing the best results from a molding sand. Some properties of chamotte are given in Table LIV. Chamotte has a rather high rate of heat transference. This high thermal conductivity is due to two factors: (1) chamotte as a calcined clay has a higher thermal conductivity than silica sand, (2) the large interstices offer a means of rapid heat transference by radiation.

When the chamotte mold has been hard rammed, the mold cavity is

TABLE LIV.—PROPERTIES OF CHAMOTTE MOLDING MATERIAL
Briggs and Gezelius[14]

	Raw-clay content, 13.3 per cent	
	Moisture content	
	5 per cent	7 per cent
Permeability		
Green..........................	2,450	2,000
Dry.............................	Unobtainable	Unobtainable
Compression strength, p.s.i., green.....	15.3	16.1
Shear strength, p.s.i.		
Green..........................	1.0	1.3
Dry.............................	5.6	6.3
Tensile strength, p.s.i., dry...........	Nil	0.3

washed with successive applications of increasingly finer chamotte paints. The mold is then dried by ovens or by torches at temperatures up to 700°-F. Further applications of mold wash may be applied after mold drying, while the mold is cooling. In some cases a less permeable material consisting of a mixture of silica sand and chamotte is used as a thin facing at the mold cavity. The backing sand consists of reclaimed chamotte.

Because of the higher calcining temperatures and the relatively high mold-drying temperatures employed, all the volatile matter is removed from the mold; and, because of the high permeability, casting porosity due to mold conditions is not experienced. Furthermore, molds of chamotte show little erosion by the molten metal, and a very minimum amount of dirt enters the metal because of mold conditions. Castings made in chamotte strip easily and cleanly. The disadvantage of its use is the necessity of training molders in an entirely new technique.

Cement-bonded Molding Sand.—Washed and graded silica sand is bonded with cement to produce air-dried molds of unusual properties and possibilities. The method is used for small castings as well as for large castings weighing several tons. Cement-bonded molding sands have frequently been described in this country under the term "Randupson process."

Only new, washed, and well-graded silica sand should be used for facing sand. The sand should be mixed with about 10 per cent cement, and a total of 4 to 5 per cent water by weight of the dry materials. This damp mixture should be rammed in the usual manner. It can be backed up with a mixture composed of cement and a lower grade of sand. Such a facing mixture, when made with high early-strength Portland cement

and protected from loss of moisture for the first 24 hr., has excellent strength and permeability for 2 or 3 days after molding, even though dried rapidly following the initial 24-hr. moist period. If normal Portland cement is used, the green strength will be lower and a longer curing period and greater age will be necessary to obtain the same set strength.

The facing sand may be mixed and rammed in the usual manner, but extended mulling should be avoided in order to reduce temperature rise and moisture loss. Mixtures at normal temperature of 70 to 80°F. should be rammed within 2½ hr. after mixing if made with normal Portland cement. Mixtures at higher temperatures, 100 to 110°F., should be

TABLE LV.—EFFECT OF AIR-DRYING TIME ON 10 PER CENT PORTLAND CEMENT
AFA 63 WASHED SILICA SAND, AT 7.0 PER CENT MOISTURE
Briggs and Gezelius[14]

Test	Hours air-dried				
	0	24	48	72	100
Permeability..........................	87	97	102	102	100
Compression, p.s.i....................	1.4				
Tensile, p.s.i.........................	Nil	0.92	1.53	1.93	1.59
Shear, p.s.i..........................	Nil	47.7	62.5	63.2	66.4

TABLE LVI.—STRENGTH AND PERMEABILITY OF CEMENT-FOUNDRY-SAND MIXES AS
INFLUENCED BY TYPE AND QUANTITY OF CEMENT
Menzel[13]

Air storage, days	Test cylinders cured at 75°F.; cement content, 12.5 per cent				Moisture content, 5 per cent by weight; cement content, 8.3 per cent			
	Permeability index No.		Compression str., p.s.i.		Permeability index No.		Compression str., p.s.i.	
	Green	Set	Green	Set	Green	Set	Green	Set
High early-strength cement								
0	112	106	4.0	930	131	122	2.6	660
1	112	107	4.0	1,120	131	128	2.6	830
2	112	112	4.0	1,005	131	136	2.6	780
3	112	118	4.0	560	131	127	2.6	450
Normal-strength cement								
0	120	103	2.9	690	141	138	1.8	415
1	120	100	2.9	880	141	140	1.8	525
2	120	107	2.9	760	141	145	1.8	470
3	120	108	2.9	360	141	143	1.8	230

rammed as soon as possible after mixing, but not later than 1 hr. The molds and cores are usually allowed to air-dry for approximately 3 days before assembling. In Tables LV and LVI are listed some properties of cement-bonded sand.

Investment Molding.—The production program for the Second World War produced a need for a method of casting precision metal parts of intricate shape that would require a minimum of machining. Prior to the war, investment molding ("lost wax" process) had been used in the production of dental and medical appliances and of some cast jewelry. During the war it was adapted successfully in the production of many aircraft and armament parts.

The investment-molding process is used for all types of metal. Steel is successfully cast by this process, although only a few steel foundries have entered this field. Most of the steel castings made by this process have been produced by new divisions of companies interested in

TABLE LVII.—PATTERN WAX AND INVESTMENT FORMULAS
Neiman[38]

	Per cent		Per cent
Wax No. 1		Investment-Prange formula[39]	
Diglycol sterate S	16.5		
Acrawax B	19.5	Silica	67.0
Beeswax	22.0		
Ceresin	42.0	Liquid { tetraethyl silicate, 8 vol. / water, 1 vol. / alcohol, 1–2 vol. / hydrochloric acid, few drops }	33.0
Wax No. 2			
Paraffin	10.0		
Carnauba	40.0		
Beeswax	10.0		
Resin	40.0		
Wax No. 3		Investment-Prosen formula[40]	
Ozokerite	61.0		
Paraffin	29.0	Silica	90.0
Resin	10.0	Magnesia	6.0
		Monobasic ammonium phosphate	3.0
Wax No. 4		Monobasic sodium phosphate	1.0
Ceresin	70.0	Liquid—water or 10 per cent hydro-	
Beeswax	30.0	chloric or nitric acid	
Wax No. 5			
Beeswax	67.0		
Resin	33.0		
Venice turpentine	(Small amt.)		

the precision casting of all types of metal. Only steel castings weighing from a few ounces to a couple of pounds have been produced by the lost wax-investment molding process.

A master mold must be prepared of the part to be cast. This mold is usually made from a low-temperature alloy or machined from brass. An increase of 1.5 per cent in mold-cavity size is necessary, to provide for wax and investment expansion. The pattern used for each casting is made by pouring wax into the master mold. Some waxes used in precision casting are enumerated in Table LVII. In some cases plastics or low-melting-temperature alloys are used as pattern material in the place of wax. After the wax pattern is made, it is surrounded or covered with an investment material. The molds are vibrated, thus allowing the investment to pack uniformly. Investments vary widely in composition, but all consist of a refractory aggregate (such as silica flour) and a binder.

Since the metal is forced into the mold under pressure or centrifugal force, the investment may be dense and thus present a hard, smooth surface. The permeability of the investment mix is very low. After the investment has hardened, the mold is heated and the wax pattern is melted and poured from the sprue. The mold is then placed in a centrifugal casting machine or in an air-pressure casting machine and poured.

The percentage of useful castings depends upon the tolerance permissible. Neiman[38] reports that on a production run of precision castings, 89 per cent were plus or minus 0.0005 in.

Slip Coating.—Some consideration is being given to the application of a clay slurry as a method of rebonding additions to molding sands.[41,42] The conversion of the clay-type bond into a true colloidal "casting slip" permits the complete covering of each particle of sand with a thin layer of clay within a brief mixing period and without introducing an excess of water.

If raw clays are used, they can be made more suitable for the purpose by the addition of flocculants or deflocculants, as required. By the use of flocculants the green strength of a sand mix can be increased materially, substantially above those values obtained with the mulling of a dry mix to which water is added. Control of the moisture content and bond strength of molding sand is more uniform when slurries are used.

A further step in the use of ceramic slips is foreseen by the author. Instead of using the slip to bond the sand, it is suggested that the entire mold be prepared as a slip or slurry. Vibration of the mold during the entry of the slip would produce a uniformly dense mold without the necessity of ramming. Permeability of the mold could be controlled by the particle size of the refractory material. A calcined clay could possibly be used in place of sand. After the mold has set patterns could be drawn from the mold through the application of heat or cold. Fast-setting mate-

rials could be incorporated in the slip.   Experiments are proceeding along these lines at the present time.

Ramming Methods.—There are a number of methods that are used in the foundry for ramming up molds.   These methods will be touched upon briefly.   First, it should be stated that in order for a mold to be properly rammed up, the sand should be thoroughly mixed.   The use of synthetic molding sands necessitates uniform distribution of the bond throughout the mass of sand and the coating of each sand grain with clay.   This requires mechanical mixing.   The batch-type muller, with its concentrated mulling or rubbing action, is used most widely and successfully for preparing rebonded or synthetic sand.   The intensive mulling and rubbing action coats each silica grain with its film of clay and develops the maximum strength of the bond used.

Less bonding material is used for a given bond strength when the sand mixture is prepared in a batch-type muller than when mixed by any other method.   The less bond required to develop a given bond strength, the higher will be the resulting permeability and flowability and the costs will lower correspondingly.   All molding operations require the ramming of sand into an irregular-shaped space, with the object of producing a uniform degree of ramming on all the pattern surfaces as well as throughout the body of the mold.

Hand Ramming.—Hand ramming consists of the ramming of small quantities of sand at a time with a rammer of small surface area.   The main advantages of hand ramming are that sand can be uniformly rammed into restricted spaces; there is no cost or upkeep of molding machines; and the method is flexible and portable.   The disadvantages of hand ramming are numerous.   The degree to which a mold is uniformly rammed by hand depends entirely upon the care given to the operation by the molder, since he can control the ramming force and the number of blows given to any particular area of the mold.   If compressed-air-hammer ramming is employed, the ramming force is fairly well controlled, although the amount of ramming given to any single area is a prerogative of the molder.   A lack of uniformity in ramming may result in casting defects caused by mold conditions such as scabs, swells, and wavy casting surfaces.   The hand rammer also tends to produce small partings in the sand, resulting in variations in the mold hardness from below the rammer face to the position of the rammer face upon subsequent blows.   From the standpoint of production, hand ramming is laborious and slow, although it is speeded up considerably by use of air-hammer rammers.

Jolt Ramming.—The process of jolt ramming consists of filling the flask with sand, raising the mold above a plate, and then permitting the mold to fall under gravity to the plate.   The impact of the mold on the plate causes the packing of the loose sand.

Jolt ramming is fairly efficient on horizontal pattern surfaces, and packing is fairly uniform. However, the effectiveness of jolting becomes less as the angle of the pattern changes toward the vertical. Patterns containing wells and pockets are not successfully and uniformly rammed by jolt ramming. In many cases such as the molding of certain valve bodies, it is necessary to hand-ram with an air hammer after the jolting operation, in order that the mold be more closely packed. Sands should have good flowability characteristics for jolt ramming.

The jolt method of ramming tends to give a greater mold hardness in the mold as the interface of mold and machine plate is approached. For this reason, jolting is particularly well suited to small, flat patterns that do not extend far from the parting surface.

Squeezing.—Generally, the practice of squeeze ramming is applied to molds made in the small-sized flasks, because of the high total pressure required. When the force is applied from the top of the sand-filled flask, the degree of packing tends to decrease in a downward direction, resulting in soft ramming, particularly in small pockets in the pattern face. The best results are obtained on light, flat castings that can be made in shallow flasks. The squeezing action is not uniformly applied to flasks containing bars or ribs.

A combination of jolting and squeezing is applied on some work, especially where even ramming on the mold face is obtained by jolting and the tendency to softness in the upper part of the mold is corrected by squeezing after the jolting is complete. In this way a fairly uniform degree of ramming can be obtained throughout a flask that does not exceed a depth of 7 or 8 in.

Sand Slinger.—The sand slinger is a mechanical means of throwing sand with a uniform force into the mold. The sand is packed through the impact of the sand itself moving at high velocity. This sand velocity is such that impact with the sand already in the flask actually permits penetration of the sand and packs in when no longer able to penetrate the underlying layer.

A great advantage of sand-slinger ramming is that the size of the mold does not influence the uniformity of ramming. Thus, while the sand slinger is for the most part used in the ramming of large molds, it can also be used on smaller molds equally well. Another advantage of the sand-slinger ramming is the aeration that goes on during the process.[43] Aeration has a beneficial effect on any molding sand, and particularly if the primary preparation is not what it should be. More uniform mold surfaces, of a greater hardness and better permeability for equal mold hardness, are obtained by sand-slinger ramming as compared with other ramming methods.

Comparison of Ramming Methods.—Studies have been made by Buchanan[43] on a comparison of ramming methods.   A flask 12 in. square and 9 in. deep was rammed by the different methods.   The size of flask chosen was not equally suitable for all types of ramming.   The sand slinger did not show its best results because the flask rammed up so rapidly that the operator had little time for maneuvering the rammer head.   On the other hand, the flask size was practically ideal for ramming by jolting. The results, illustrated in Fig. 198, show a series of hardness tests made at various levels from top to bottom of the flask, as rammed by various means.   Considerable variation between top and bottom of the mold is shown with jolt and squeeze ramming.
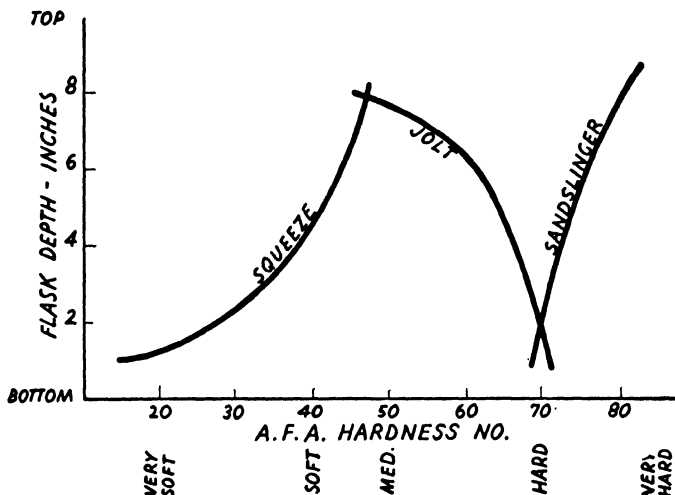
Fig. 198.—Mold hardness as shown by mold-ramming method: (1) Sand slinger, (2) jolting, (3) squeezing.   (*Buchanan.*[43])

Drying Out of Sands.—Water used for the tempering of synthetic sands evaporates readily, probably because of the permeable nature of the sands, the small amount of clay, and the minimum of fines.   The loss of moisture is more apparent and more of a problem because of the small amount of water present in synthetic-bonded sands.

The present practice in steel foundries using synthetic-bonded sands is to make allowance for probable evaporation by adding more water to the sand on hot days.   Some foundries have attempted to reduce moisture loss by the addition of certain fluids to the sand, such as glycerin and fuel oil.

Studies on the effectiveness of certain added materials in reducing the drying out of synthetic-bonded sands have been made by Dunbeck.[44]

He observed that lithium chloride additions are about the most effective agent in decreasing the drying out of the sands. However, for the lithium chloride to be effective in decreasing the drying-out properties, it should be present in quantities of about 10 per cent of the mix. A 2 per cent addition has but little effect in preventing the sand from drying out. In most cases, 10 per cent addition of lithium chloride causes a drop in the green, dry, and hot strength of the sand.

Cereal binders are capable of reducing the rate of drying out of fire clay and southern-bentonite-bonded sand, but they are less effective with western-bentonite-bonded sand. A comparatively small reduction in the rate of drying out is of appreciable benefit.

Fuel-oil additions do not appreciably reduce the rate of drying out. Sands treated with fuel oil feel wetter than is actually the case and are more workable. Since the chief disadvantage of dry sand is that it is brittle, additions of oil may be of value in that they increase resilience and workability. The principal objection to the use of fuel oil is the annoying smoke that is created upon pouring. Also, increased quantities of clay are required on rebonding.

Hot Sand.—One of the problems growing out of continuous foundry operations is that of maintaining the sand at a temperature suitable for molding. Hot sand is responsible for the moisture's drying out too rapidly. With rapidly varying water contents, the sand will not exhibit uniform strength or other properties. Hot sand is likely to cause sweating on the patterns and drops in copes.

Evaporation of moisture in the sand after shakeout is one means of cooling sand for continuous-high-production units.[45,46] When the moisture drops below a certain point, the sand in the conditioning unit gradually increases in temperature, and this rate of increase is directly proportional to the moisture deficiency after shakeout. The remedy is to add more moisture to the returning hot sand. This may be accomplished by the correct water addition to the sand while it is traveling on an exhausted conveyer; or moisture may be increased by allowing more green sand to overflow to the hot sand-return belt where it combines with the shakeout sand and furnishes the necessary moisture for evaporation.

Sand may be cooled by dropping it in a light shower onto baffles, producing a cascade effect. If air is forced through the baffles, cooling and removal of fines can be accomplished. Sand may be cooled during mulling if modern mullers with forced air drafts are used.

PROPERTIES OF CORES

There are two major classes of cores: (1) green and (2) dry. Green-sand cores are similar in composition to green molding sand, and their properties can be controlled in the same manner as is illustrated under

the properties of mold sand. Dry cores have little strength, rigidity, or hardness in the green condition, and they develop these properties only after they are dry. In most cases heat is applied to dry cores; however, there is a class of air-drying cores.

Cores have been classified by Dietert[47] according to the method in which the core binder stiffens or hardens the cores. There are five main types, based on binder characteristics as follows:

1. Cores that harden at about 70°F.
    a. Sodium silicate
    b. Portland cement
    c. Rubber cement
    d. Silicon esters
2. Cores that dry on heating
    a. Vegetable oils
    b. Polymers, either mineral or vegetable
    c. Marine-animal oil
3. Cores that harden on cooling after being heated
    a. Gum resins
    b. Resins, wood
    c. Resins, coal tar
    d. Pitch
    e. Resin oils
4. Core binders that adhere on heating
    a. Sulphate
    b. Proteins
    c. Cereal
5. Core binders that are of the clay type
    a. Fire clay
    b. Bentonite

The various types of binder as listed above are supposed to confer upon the core certain desired properties for a particular use. However, in most cases the fields of usefulness overlap considerably; and with a certain amount of manipulation it is possible to use one binder, or a combination of binders, to fulfill nearly all needs.

The requirements that a binder should possess to produce good cores are similar to those for the bonding of molding sand; *i.e.*, they must confer strength, flowability, and surface hardness and must generate a minimum of gas. The bond should also be of such a character that it will permit a core to collapse after the metal has solidified and will not cause the core to distort during baking or cause sand to stick in core boxes. Still another important requirement is that the binder should resist, as much

as possible, the absorption of moisture by the core after it is baked or dried.

Types 1, 3, 4, and 5 as listed above will not be discussed in this section, because when these binders are used singly, they give conditions similar to those reported on pages 343–383. In some instances the binders of the above named types are used with the oil binders in core mixes. Information available on such combinations will be presented.

Core Oils.—Oils used for cores are composed of drying or semidrying oils, vegetable, mineral, and marine-animal oils. Proprietary core oils consist of perhaps several different oils with other dissolved substances such as resin. Core oils have been adjudged according to the following properties:

Acid number
Saponification number
Iodine number
Flash point
Fire point
Viscosity
Specific gravity
Color
Stickiness
Green strength
Green permeability
Flowability
Baked permeability
Baked strength
Core hardness
Collapsibility
Gas content

In discussing the properties of core oils, typical core oils as used in steel foundries will be reported upon.

Chemical and Physical Properties of Core Oils.—Core oils having a high acid value (over 10) indicate a high deflection in the baked core. Low acid value in the core oil seems to increase the flowability of the green-oil-sand mixture. No other property of the green-sand mixture appears to be affected by the acid value of an oil. The acid value of a core oil does not indicate the stickiness of the oil. Resin in a core oil can usually be detected by a high acid value of the core oil. Fish oil is not necessarily indicated by a high acid value. The acid value of some core oils is shown in Table LVIII.

Core oils that have saponification values over 130 develop the highest maximum core hardness and tend to retain this hardness when subjected to humid atmospheric conditions. Core oils that have the higher saponifi-

cation values have a tendency to bake out more completely, as indicated by the presence of less core gas.   Practically all dark-colored core oils have saponification values below 110.   For many oils the specific gravity of the oil tends to decrease as the saponification value of the oil increases.

TABLE LVIII.—CHEMICAL PROPERTIES OF SOME STEEL FOUNDRY CORE OILS

Type	Acid No.	Saponifi-cation No.	Iodine No.
1. Raw linseed...	3.8	188	141
2. Boiled linseed.......................................	5.3	197	137.3
3. Linseed + fish + light mineral oil + resin*			
Oil a†...	19.5	182	122
b...	29.8	155	92.9
c...	32.6	156	102
d...	39.4	130	84.7
4. Linseed + light mineral oil + resin*.................	42.1	156	94.2
5. Linseed + fish oil + solvent*........................	4.6	115.5	63.1
6. Linseed oil + mineral polymers*			
Oil a...	7.1	89.6	117.8
b...	2.3	78.2	107.9
c...	7.2	104	95.5
7. Mineral-polymer type*			
Oil a..	10.5	68.8	109.1
b...	5.3	83.8	110.8
8. Mineral polymers, with resin and vegetable oil*			
Oil a...	42.0	97.2	97.6
b...	25.4	103	123
9. Light mineral oil + resin*..........................	68.6	90	123

* Percentage composition of oils not determined.

† Trade names of oils not given.

The iodine value of a core oil seems to bear little relationship to the green or baked physical properties of the oil sand, and practically no relationship to the color, flash or fire points, specific viscosity, or specific gravity of the oil.   Core oils that have a low iodine value and a high acid value generally exhibit a high rate of core collapsibility, especially at temperatures above 1500°F.   A comparison of the iodine value, saponification number, and acid number of some core oils is shown in Table LVIII.

The flash and fire points of nearly all core oils are lower than those of either the raw or boiled linseed oils.   Most proprietary core oils contain oils other than linseed, which give to the core oil a low flash and fire point (Table LIX).

The average specific gravity for light-colored oils is less than it is for the dark-colored oils.   The dark-colored oils often have high specific-viscosity values, and in some cases are responsible for increasing the

stickiness of the green sand in core boxes. Oils that are dark usually contain mineral polymers, as is shown in Table LIX.

Green Mechanical Properties of Cores.—The green strength of cores bonded only with core oils is very low. It is so low, in fact, that cores prepared of sand, oil, and water must be handled with exceeding care. The cores are usually made in halves and carefully removed from the core box onto plates that are placed directly into core ovens. After the cores are baked, they are of sufficient rigidity and strength to permit the two core halves to be pasted together.

TABLE LIX.—PHYSICAL PROPERTIES OF SOME CORE OILS

Type	Flash point, °F.	Fire point, °F.	Specific gravity, 60°F.	Viscosity, 60°F.	Color	Stickiness
1. Raw linseed	570	665	0 932	4 4	Light	Medium
2. Boiled linseed	560	660	0 938	5.7	Light	Medium
3. Linseed + fish + light mineral oil + resin*						
Oil a†	315	535	0.944	6 7	Light	Slight to medium
b.	185	250	0.943	6 8	Light	Slight to medium
c.	185	220	0 943	4.8	Light	Medium
d.	180	210	0 943	6.8	Light	Medium
4. Linseed + light mineral oil + resin*.	140	165	0 938	7 6	Light	Medium
5. Linseed + fish oil + solvent*.	...	...	0.961	3.4	Cream white	Heavy
6. Linseed oil + mineral polymers*						
Oil a	225	260	0 945	8 4	Dark	Medium to heavy
b.	220	245	0 946	9.4	Dark	Medium
c.	215	275	0 958	5 9	Dark	Medium
7. Mineral polymers*						
Oil a	215	230	0.945	8.7	Dark	Slight
b.	185	205	0.958	9 5	Dark	Medium
8. Mineral polymers + resin + vegetable oil*						
Oil a	180	195	0.954	19 3	Dark	Heavy
b.	195	225	0.943	8 6	Dark	Medium
9. Light mineral oil + resin*.	155	165	0 922	3 6	Light	Medium

* Percentage composition of oils not determined.
† Trade names of oils not given.

If it is necessary that cores be given a strength greater than can be obtained by oil and water alone, other bonding materials are added. Cereal binder, in additions of about 1 per cent, will double the green strength of clay-free mixtures. Bentonite or clay additions will increase the green strength to values comparable with that found in molding sands. The green strength of a sand-oil-water mix can also be raised by increasing the fineness of the sand and the moisture content. However, stickiness also increases with moisture increase. Core baking is slowed up with increasing moisture content.

It is normal practice in production to use a sand-oil ratio of about 80 to 1 by volume. A ratio for standard testing of 76 to 1 by weight (approximately 80 to 1 by volume) is often used. Other ratios, such as 100 to 1 by weight, are also used.

The green properties of core sands of a sand-oil-water mix are given in Table LX. The standard AFA washed silica sand of a sieve classification of 50 to 70 mesh was used. It will be noted that regardless of the type of core oil used, the green strength is about 0.30 p.s.i. The flowability value is excellent with all oils and does not vary much. The greatest change is found in permeability, with values from 160 to 200. In general, the toughness of the green-sand mixtures tends to decrease with the increased specific gravity of the oil. Also, as the toughness of an oil-sand mixture increases, the stickiness of the green sand in the core box becomes less.

TABLE LX.—GREEN PROPERTIES OF 50- TO 70-MESH WASHED SILICA SAND WITH VARIOUS CORE OILS AT 0.76* PER CENT

Type	Moisture, per cent	Permeability No.	Compression str., p.s.i.	Flowability No.	Toughness No.
1. Raw linseed..........................	3.5	159	0.35	93	21.0
2. Boiled linseed........................	3.7	168	0.31	90	18.9
3. Linseed + fish + light mineral oil + resin†					
Oil a‡...........................	3.6	176	0.30	91	20.4
b.............................	3.6	189	0.37	90	18.5
c.............................	3.8	190	0.31	91	16.7
d.............................	3.7	182	0.36	91	17.6
4. Linseed + light mineral oil + resin*..	3.6	177	0.32	90	19.2
5. Linseed + fish oil + solvent†.........	3.5	179	0.29	93	11.3
6. Linseed oil + mineral polymers†					
Oil a............................	3.6	158	0.39	89	14.0
b.............................	3.5	190	0.32	93	14.7
c.............................	3.6	180	0.35	91	14.0
7. Mineral polymers†					
Oil a............................	3.8	204	0.37	91	18.5
b.............................	3.7	170	0.44	93	15.0
8. Mineral polymers + resin + vegetable oil†					
Oil a............................	3.5	198	0.33	90	12.2
b.............................	3.6	192	0.29	92	17.7
9. Light mineral oil + resin†............	3.8	197	0.30	89	17.7

* 0.76 per cent approximate equivalent to 80 to 1 sand-oil volume ratio.
† Percentage composition of oils not determined.
‡ Trade names of oils not given.

Baked Properties of Cores.—Cores made with bentonite or fire clay or cereal binders, or any combination of these, will have properties similar to those of dried molding sands. Cereal binders are added with core oil to cores, to increase their strength and assist in the prevention of metal erosion. Cores containing cereal binders alone usually do not have sufficient strength for most core applications. If bentonite or clay is used alone, the core burns out too slowly and produces cracked castings in many cases.

The use of core oil alone produces cores of good strength and fast collapsibility. If the strength is not sufficient for the particular application or if collapsibility is too fast, it is advisable that a combination of bentonite or clay and core oil be used to bond cores.

The baked properties of cores using various types of core oil and a standard AFA sand of 50 to 70 mesh are given in Table LXI. Core-oil additions were made at 0.76 per cent. It will be observed that the core

TABLE LXI.—BAKED PROPERTIES OF 50- TO 70-MESH WASHED SILICA SAND WITH VARIOUS CORE-OIL ADDITIONS AT 0.76 PER CENT AND 3.5 PER CENT MOISTURE

Type	¾-hr. baking time				1½-hr. baking time			
	Best temperature,°F.	Max. tensile str., p.s.i.	Hardness No.	Permeability No.	Best temperature,°F.	Max. tensile str., p.s.i.	Hardness No.	Permeability No.
1. Raw linseed................	450	135 8	71	160	425	132.5	67	168
2. Boiled linseed...............	450	153 7	71	160	400	118.2	71	177
3. Linseed + fish + light mineral oil + resin*								
Oil a†	450	133.5	61	175	425	139.0	63	192
b....................	450	119 7	57	179	400	165.2	65	193
c....................	425	113.3	62	181	400	116.3	61	190
d....................	425	109.7	58	177	400	118.2	59	192
4. Linseed + light mineral oil + resin*................	450	87.7	51	172	400	118.8	59	188
5. Linseed + fish oil + solvent*.	450	84.2	50	177	400	68.7	50	190
6. Linseed oil + mineral polymers*								
Oil a.....................	450	134 5	60	169	400	98.9	56	168
b.....................	450	128.7	58	177	400	110.3	57	189
c.....................	450	148.0	64	169	425	146.8	58	188
7. Mineral polymers*								
Oil a.....................	450	77.7	43	168	400	77.3	48	201
b.....................	450	105.7	54	178	400	109.2	55	188
8. Mineral polymers + resin + vegetable oil*								
Oil a.....................	450	99.4	44	177	425	80.2	47	194
b.....................	450	70.3	44	190	400	119.0	55	205
9. Light mineral oil + resin*....	350	10.0	0	...	350	10	0	205

* Percentage composition of oils not determined.
† Trade names of oils not given.

oils containing an appreciable amount of linseed oil have the best baked
properties.   The light-colored core oils generally have better baked prop-
erties than do the dark-colored core oils (compare with Table LIX).   The
dark oils listed in the table do not have high resistance to high baking
temperatures.   Core oils that have the highest baked strength also show
a high core-hardness value.   Apparently, the core hardness increases
with the tensile strength of the core.   Oils high in linseed oil also develop
an early strength in short heating time.

It has been reported by Casberg and Schubert[48] that the ratio of the
saponification number (or iodine number) to tensile strength for the oils
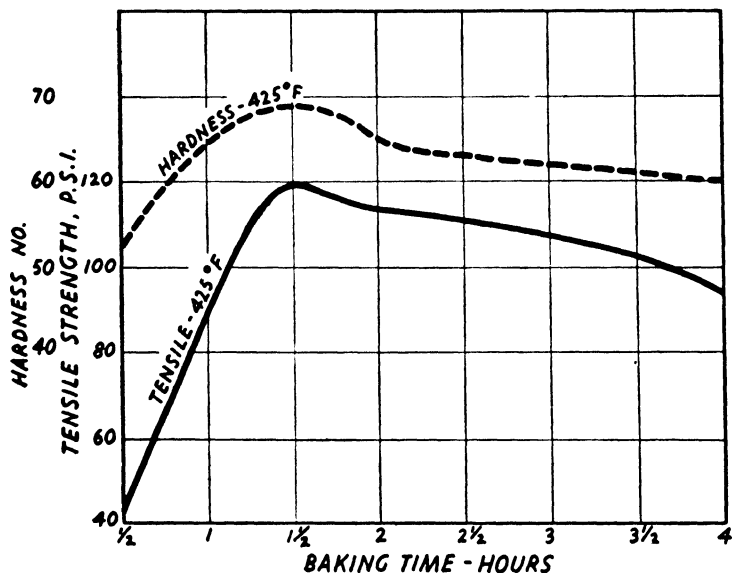
Fig. 199.—Effect of baking time at 425°F. on the tensile strength and hardness of 0.76
per cent raw linseed oil bonded cores: AFA standard 50- to 70-mesh sand; moisture 3.5 per
cent.   (*Woodliff*.[51])

that they tested can be represented by a straight-line equation.   A com-
parison of Tables LIX and LXI will show that no correlation such as
obtained by Casberg and Schubert was possible.   A comparison of the
properties of the core oils of Table LXI indicates that the short baking
cycle requires an increase in baking temperature of approximately 10 per
cent to obtain the average baked properties of the longer baking cycle.

Petroleum extracts can replace up to 40 per cent of linseed oil, accord-
ing to Davies and Rees,[49] without adversely affecting the tensile strength
of sand mixes bonded with 2 per cent oil.   If more than 50 per cent of
the linseed oil is replaced by extract, the tensile strength falls off sharply.

The temperature at which an oil will develop a maximum core hard-
ness will depend upon the characteristics of the oil.   However, with

nearly all oils the maximum core hardness is developed between the temperatures of 350 and 425°F. In many cases the maximum core hardness is reached at a lower temperature than that which gives the maximum tensile strength. For some oils, the maximum core hardness and tensile strength occur at the same temperature. To develop the maximum hardness in cores, they should not be baked at temperatures of more than 50°F. above the temperature that gives the maximum core hardness. At higher temperatures the hardness falls off rapidly.

The length of baking time at the maximum core-hardness temperature affects the hardness and strength of cores (Fig. 199). Some oils, espe-
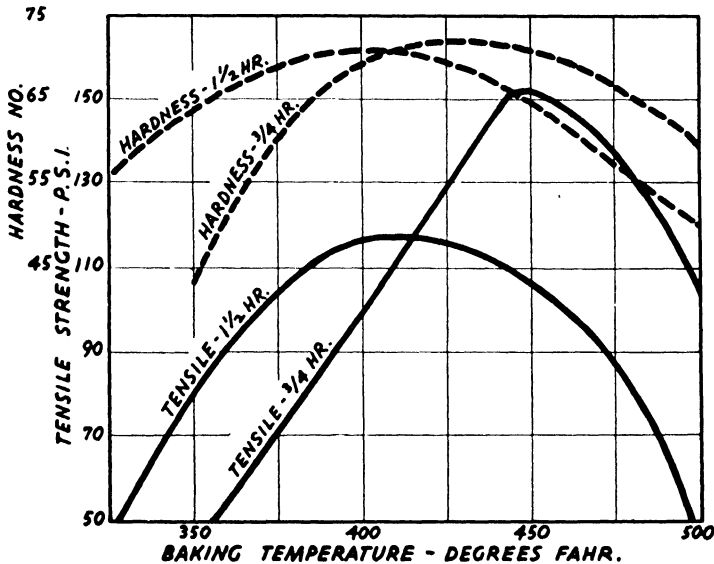
Fig. 200.—Effect of baking temperature at two baking times on tensile strength and hardness of 0.76 per cent boiled linseed oil bonded cores; AFA standard 50- to 70-mesh sand; moisture 3.7 per cent. (*Woodliff.*[51])

cially those high in linseed oil, develop hardnesses approaching the maximum hardness in short baking times. The length of baking time on the temperature at which a core is baked affects both the tensile strength and hardness of the core (Fig. 200). Increasing the oil in a core-sand mixture increases the strength and hardness,[50] as shown in Fig. 201. The hardness increases in a straight-line function, while the strength increases in a parabolic function. Increasing the amount of oil decreases the permeability slightly. It will, however, materially increase the amount of gas generated.

The baking times, as shown in Figs. 199, 200, and 201, cannot be used as criteria for baking cores in the foundry, because the size of the core will control the baking time. Laboratory ovens are more efficiently con-

trolled as to air circulation and to temperature. The graphs, however, show clearly that the length of baking time materially affects both the core hardness and tensile strength.
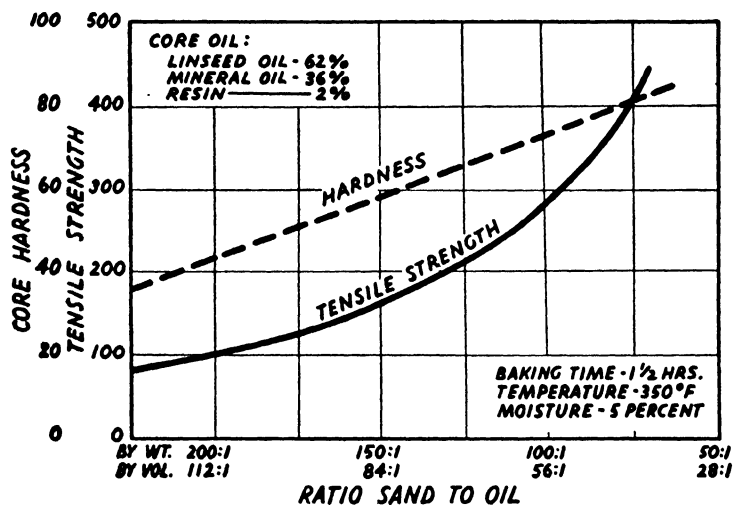
FIG. 201.—Properties of baked cores when sand-oil ratio is varied. (*Dietert and Woodliff.*[20])

Baked cores that stand after baking in humid atmosphere lose a proportion of their strength and hardness. In order that an estimate can be made of this loss, the baked-core test specimen is stored after cooling, in a saturated humid atmosphere for a timed period. The behavior of core oils when subjected to a humidity test seems to indicate that all types

TABLE LXII.—LOSS IN CORE STRENGTH AND HARDNESS IN HUMID ATMOSPHERE

Type	Per cent loss in saturated atmosphere for 3 hr.	
	Hardness	Tensile str., p.s.i.
Raw linseed............................	15.8	49.2
Boiled linseed...........................	10.5	56.7
Linseed + fish + light mineral oil + resin....	13.3	39.8
Linseed + light mineral oil + resin..........	19.8	46.1
Linseed + fish + solvent...................	27.7	40.2
Linseed oil + mineral polymers.............	14.5	45.5
Mineral polymers.........................	17.7	45.3
Mineral polymers + vegetable + resin........	22.5	48.1
Light mineral oil + resin...................	Too weak for test	

of core are much weakened in a humid atmosphere, and their loss in baked strength is high. The percentage of loss in tensile strength for any one time period is much greater than the percentage of loss in hardness, as is shown in Table LXII. The core oils that maintain a high percentage of their original core hardness are those containing a high percentage of vegetable oil.

High-temperature properties of cores are considered in Chap. XI.

Core Production.—Core production is an item of major importance in the steel foundry. In fact, many foundries found that lack of core-department facilities was the bottleneck of casting production during the first months of the Second World War. A considerable amount of manually made cores is produced in the steel foundry today. These cores are for the most part very small or rather large cores. Nearly all of the other-sized cores are machine made. Machine molding of cores is similar to machine molding of molds, with the exception of core blowing. The making of cores by blowing the sand into the core box under pressure is a practice that has been well developed during the past few years. The method permits the rapid production of small- to medium-sized cores of all stages of complexity.

An assembly of cores may constitute the entire mold. Such molds may vary in size from small stack molds to large pit molds. Stack molding is an interesting development wherein cores are stacked on top of each other to form castings run from a common gate.

Considerable thought is required in the building of some cores. In order to prevent the core from sagging or from being buoyed up by the metal, it is necessary that the cores be built around a frame of rods tied together or around a specially constructed arbor. These arbors are often constructed of cast iron in such a manner that they can be broken when the casting has solidified, to permit the casting to contract without being hindered by the core construction.[51]

In certain cases, because of the development of hindered contraction, it is advantageous to construct the top half of the core from green sand. At other times, the center of the core is made hollow or filled with easily compressible materials. A discussion of modern core practice and techniques is covered by Dietert.[47]

RECLAMATION

Molten steel poured into sand molds will result in the splitting of some of the sand grains, deterioration of some of the clay bond, and the burning out of a certain amount of the organic bonding material. A quantity of the sand grains will be coated with iron oxide products. The extent to which the deterioration of the molding and core sand takes place depends upon the degree of heat penetration into the mold. A full discussion of

the effect of molten steel on the properties of molds and cores is given in Chap. XI.

If a molding sand is reused repeatedly with a little clay addition made prior to each reuse, it shortly becomes so low in permeability, deformation, and toughness that it is not advisable to use it further, as defects caused by mold conditions will increase out of all proportion to what would be found if new sand were used. It is a matter of record that if no method of sand reclamation is used in the steel foundry, the sand consumption will be from 1 to 1.5 tons of sand for each ton of steel castings produced.

To reclaim a molding sand, it is necessary to remove the dust and very fine particles in the sand caused by the splitting of the sand particle and the presence of silt due to the breakdown of the clay-bonding material. It is also necessary to remove the coating of foreign substances from the sand grain and to remove this material from the sand. If these operations are successfully performed, the remaining sand will be composed primarily of quartz, which can be charged into a muller as a substitute for new sand in the making of molding-sand mixtures.

The reclamation of molding sand and cores has resulted in the reduction of new sand from the 1 to 1.5 tons of sand per ton of steel castings produced to 200 to 1,000 lb. of sand per ton of steel castings produced. The manner in which reclamation can eliminate the fines in a sand is illustrated by the following fineness test of a foundry-heap sand, made before and after reclamation:

Mesh No	20	30	40	50	70	100	140	200	270	Pan
Before reclaiming	1.0	7.7	40.5	29.5	13.1	4.5	1.5	0.5	0.5	0.7
After reclaiming	0.6	6.6	49.9	29.6	9.6	2.7	0.6	0.2	0.1	0.1

The sand grains close to the mold-metal interface will become coated with a black coating that adheres to the grain. The percentage analysis for the coating of grains is about as follows:

C	SiO_2	Al_2O_3	Fe_2O_3	CaO	MgO
9.5	60.0	25.0	2.0	3.0	0.5

The coated grains are much less refractory than new sand, having only about 2550°F. sintering point, while the new sand has a sintering point of around 2800°F.

The reclamation of all sand prior to reuse is not necessary, since only the first 1 or 2 in. surrounding the mold-metal interface are altered by the metal heat. All the backing sand, therefore, does not require reclaiming. However, a sufficient portion of it should be reclaimed to prevent the

backing sand from building up in fines and lowering the permeability. As a first step in sand reconditioning, the lumps formed by mold drying or by the heat of the metal are broken down. This operation can best be accomplished by a shakeout machine of the vibrating or reciprocating type. The separation of metallic pieces by means of a magnetic separator should then follow.

Dry Reclaiming.—In some cases revolving screens with lifting shelves have been used prior to grain-size classification, as a method of breaking down loosely bonded lumps and of creating an intimate contact between the air stream and the surfaces of the individual grains. Such an operation permits the cooling and drying of sands as well as the removal of dust. The main requisite of the operation is that there be an adequate air supply of regulated velocity.

The revolving screens are able to make a sand classification of three sizes: 1- to 3-in. lumps, 20-mesh to 1-in. material, and sand finer than 20-mesh. The lumps may again be returned to the screen or fed to a small pulverizer for reducing them to approximately 20-mesh size. Screening for grain-size classification is accomplished, generally, by vibrating, either single deck or multiple deck. It should be pointed out that, in classification by screening, the screen will reject every particle larger than its opening, though every particle smaller than the opening may not pass through it. Any screening operation can control only the maximum and minimum size of the product; this can be done reasonably well only when the feed is properly adjusted, the screen is long enough so that the sand grains may have numerous attempts to pass through it, and the screen is not clogged.

An effective way to regulate dry reclamation is the substitution of a sand-cascade unit for the vibrating screens or, in some cases, to install such a unit ahead of the vibrating screens.[26] The cascade unit is constructed of a number of louvers. The sand stream falls, bouncing from louver to louver, and in doing so is exposed to an air stream moving perpendicularly to the falling sand. The air stream removes the dust and fine particles and also acts to cool the sand. The degree of classification depends upon the adjustments made on the louvers and the air pressure. The sand-striking action on the louvers assists in the partial removal of the black coating on the grains. Dry reclamation does not have much effect on the oxide coating. A more pronounced scrubbing action is required to remove this coating; hence, all dry reclaimed sands are fairly dark colored. The removal of the fines assists in raising the sintering point, however, and such sands have found wide application as facing sand with little or no addition of new sand.

Wet Reclamation.—The wet method of reclamation has three advantages over the dry method: (1) handling wet sand will permit pumping

instead of using long belts and vertical elevators for conveying the sand; (2) the wet method eliminates the need for dust-suppression equipment; and (3) the wet scrubbing of the sand grain removes the black coating.[53,54]

In some cases the wet method of cleaning is used in connection with hydroblast cleaning, with a proportion of the sand to be reclaimed being mixed with water and forced at about 1,200 lb., pressure against castings to be cleaned.   During the operation of pressure blasting, the sand is scrubbed and a large percentage of the black coating around the grains is removed.   Sand that does not go through the pressure blast is not effectively scrubbed unless other steps are taken.   In one type of wet-reclamation system a sand-scrubber pump is provided, whereas in another system the wet sand is mulled for several minutes prior to entering the classifier system.   The sand, while being mulled, is quite wet.

Excellent classification can of course be obtained by water classification, and a sand graded better than the original sand can be prepared. The sand is dried by the use of centrifuge equipment, followed by a revolving barrel oven or by depositing the wet sand on a revolving sand drier with hot air being forced through the sand.   An example of the removal of fines on water classification is illustrated as follows:[53]

Screen	20	30	40	50	70	100	150	200	Pan
New sand	3.1	14.1	47.6	20.9	7.7	3.6	1.8	0.6	0.6
Washed used sand	3.2	10.0	35.3	26.5	15.5	6.7	1.9	0.5	0.4

Reclaimed sands that have been thoroughly scrubbed and classified will produce bonded properties similar to those of new sand.

If an efficient scrubber is placed in the system, the reclaimed sand will have a considerable portion of the black coating removed.   The sand will usually have a gray appearance, and there will be a certain amount of luster to the grains.   If all the sand placed in the wet system does not enter the scrubber, the coating will not be removed, and the reclaimed sand will have a dark appearance without luster.   Some wet reclaiming systems have not been too effective because they have not employed an efficient method of scrubbing the sand.

Thermal Reclamation.—The thermal method of sand reclamation has received considerable attention.   Prior to the advent of this method many foundrymen were of the opinion that a completely successful reclamation process must approximate new-sand results and produce a uniform sand.   It has been held that it was difficult to obtain and maintain a definite scrubbing action by wet reclaiming methods.   The thermal process, however, permitted the entire renovation of the used sand, even to the restoring of its natural color and luster.[55,56]

The essential units of the thermal sand-reclamation system are a high-temperature rotary kiln and a cooler, with essential burners, fans, and dust collector. Operation consists in reducing the sand lumps by pulverizing and screening so that a uniform delivery of sand under 20-mesh size is made to a rotary kiln operating at a calcining temperature of 1400 to 1500°F. The high temperature, along with the abrading action of the rotary kiln, removes the carbonaceous matter and oxide adhering to the sand grains, by calcining and vaporizing. The clay bond is dehydrated and calcined in the sand grain. The calcined clay is removed by dry mulling with an air blast passing through the sand. The dry mulling following the thermal treatment scorifies, smooths, and cleans the sand grains, while the particles of dehydrated clay are removed by the air stream to dust collectors. If further classification is necessary, the sand can be cascaded through an air stream. Thermally reclaimed sand will act as new sand when rebonded and will give similar properties.

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CHAPTER XI

EFFECT OF MOLTEN STEEL ON SANDS, CORES, AND WASHES

MOLDING AND CORE SANDS

The molding sand and cores forming the mold are subjected to very pronounced temperature changes when molten steel is poured into the mold. These temperature changes take place in a short time interval, and the magnitude of the changes is great. Within a time interval as short as 10 sec., a temperature change of from 70 to 2200°F. may result in the sand at the mold-metal interface.

What happens to sand that comes in contact with molten steel? Does the heat destroy the properties of the sand, or can it be used over again? Are the properties that the sand develops at high temperatures important and worthy of consideration in the production of quality-steel castings? These and similar questions will receive attention in the chapter.

Deterioration of Molding Sands in Use.—Not many commercially usable materials are acceptable for molding and at the same time are able to withstand the sudden great temperature changes. Silica sand is one of the materials that is fairly adaptable to these conditions. Sand is affected by the high temperatures and will deteriorate. This fact is fairly well known by those working with molding sands. Deterioration is the result of the change that takes place in the sand grains and in the bond.

When molten steel is poured into a mold, the sand at the mold-metal interface is heated by the molten steel. The sand grains expand with some of them splitting, and the bond changes its characteristics so that it is no longer an active bonding agent. Furthermore, the pouring of steel into sand molds introduces ferrous silicates into the sand, and these, although small in amount for one pouring, would accumulate rapidly were the sand continuously reused without reclaiming.

Sand molds, unlike metal molds, are highly permeable to atmosphere. The result is that oxygen is usually present to react with the steel at the mold-metal interface, to form iron oxide. The iron oxide in turn reacts with the silica sand to form a low-melting flux of ferrous silicate, which penetrates the sand. It has been estimated[1] that approximately 15 per cent of the scale formed at the mold face consists of FeO. The sand not only deteriorates in its refractory characteristics, but the sand grains split up into smaller particle sizes because of the heat effect of the molten metal.

400

Effect of Temperatures on Sand Grains.—Molding sand, free from bonding materials, consists almost entirely of silica in the form of the crystalline mineral quartz. Silica can form two other crystalline minerals, tridymite and cristobalite. Quartz is the form in which silica is usually found; the others are of rarer occurrence.
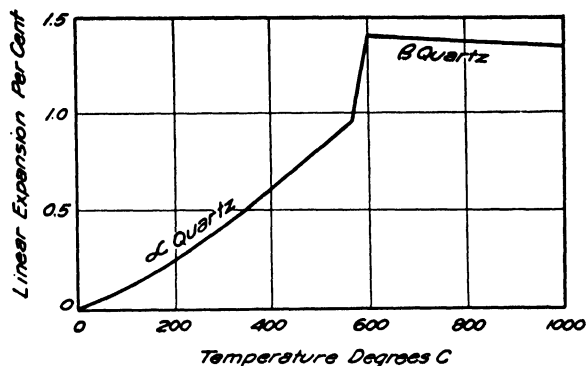
FIG. 202.—Linear expansion of quartz. (*McDowell.*[3])
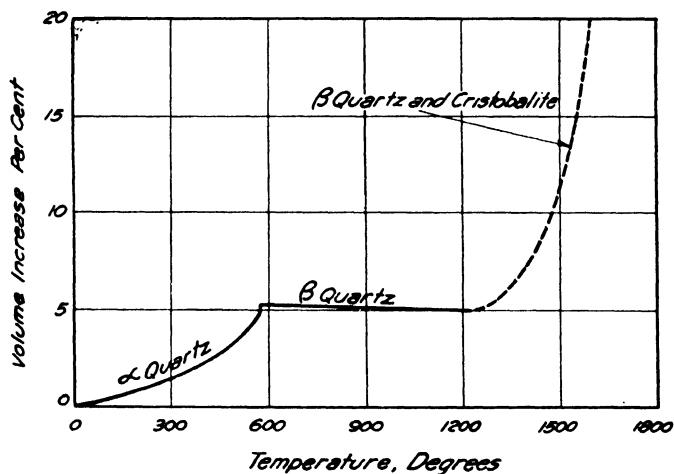
FIG. 203.—Volume increase of quartz. (*McDowell.*[3])

Quartz expands upon heating, and at a temperature of 575°C. (1067°F.) quartz changes from the alpha to the beta phase. A linear expansion of approximately 0.4 per cent accompanies the phase change (see Fig. 202). Above the phase change there is no increase in expansion until 1250°C. (2282°F.) is reached. At 870°C. (1598°F.) beta quartz is no longer stable, tridymite being the stable form of SiO_2. However, this phase change in the absence of a flux apparently does not take place, and even in the presence of a flux it takes place slowly. At 1250°C. (2282°F.) cristobalite is the stable phase. At this temperature the phase change is

extremely slow; but at higher temperatures such as 1600°C. (2912°F.), the transformation may be completed in about 1 hr., and the expansion under these conditions might be as much as 15 to 20 per cent, as shown in Fig. 203. At 1625°C. (2958°F.) cristobalite melts to silica glass. The alpha-to-beta change of quartz is reversible, whereas the inversion to cristobalite is apparently permanent and the silica does not revert to quartz on cooling.

It has been found[1] that quartz has a strong tendency to crack at the alpha-to-beta phase change. This shattering is pronounced in the larger quartz grains. Since the tridymite-to-cristobalite phase change does not take place with any appreciable speed below 1400°C. (2552°F.), this change may be dismissed as a factor of any importance in the disintegration of molding-sand grains.

Observations by Dierker[1] show that only the grains that come in contact with the metal are broken up to any noticeable extent. Since these grains absorb ferrous iron, their disintegration is desirable, for the resulting small particles can be removed in the reclaiming operation and the iron oxide content of the sand kept at a low point.

Effect of Temperature on Binders.—The value of clay bond contained in a molding sand lies in its colloidal properties. At higher temperatures the water of hydration is driven off and the colloidal properties are permanently destroyed. The most prominent types of clay binder for molding sands are fire clay and bentonite. The bonding substance in fire clays is chiefly the clay mineral kaolinite (Al_2O_3-$2SiO_2$-$2H_2O$). Kaolinite absorbs water in a reversible manner up to a temperature of 806°F. However, the reabsorption of water after a previous heating is slow. Kaolinite loses most of its water of hydration between 734 and 979°F. After being heated to 842°F., kaolinite is altered mineralogically; and at 1652°F., kaolinite eventually changes over to the mineral mullite ($3Al_2O_3$-$2SiO_2$,) and a high-temperature form of quartz (SiO_2). The material between the temperatures of 842 and 1652°F. is in the transition state. A reduction in the particle size of the kaolinite is found to take place during the transition state. A reduction to about one-fourth the size from 20 to 5 microns is recorded.

The bonding substance of bentonite is the clay mineral montmorillonite (Al_2O_3-$4SiO_2$-XH_2O). An inherent property of montmorillonite is its high absorptive ability. It absorbs water in a reversible manner up to 1022°F. Montmorillonite loses most of its water of hydration between 932 and 1112°F. The mineral loses its ability to swell at 1112°F. because of the formation of an anhydromontmorillonite. The lattice grouping breaks down at about 1832°F. The exact mineralogical changes on heating montmorillonite to the high temperatures have not been determined; however, it probably changes over to mullite.

The clay at the mold face is changed to the mullite form and agglomerates with the sand grains. Most of the clay back from the mold face that is affected by temperature is dehydrated and occurs in used sand either free or entrapped by the other binders as a nonbinding coating on the sand grains. The clay at the high temperatures breaks down into small particles possessing none of the properties of clay and remains in the molding sand as increasing quantities of silt. Thus silt may be produced in a molding sand from the bonding substance as well as from the breaking down of silica-sand particles.
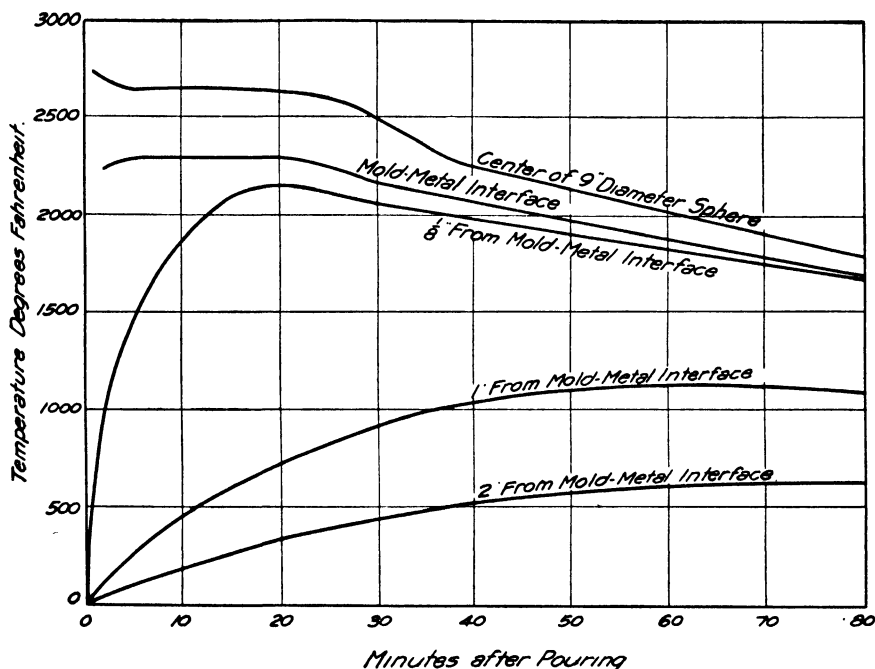
Fig. 204.—Temperature gradients in metal and mold in a 9-in. diameter sphere casting. (*Briggs and Gezelius.*[5])

Organic binders such as oils, starches, and water-soluble materials are burned out completely near the mold-metal interface. Farther away, they are merely charred, and at a greater distance they are unaffected and remain as unaffected coatings on the sand grains.

Temperature Attained by Molds.—It has been shown that sands and clays deteriorate at elevated temperatures. It now remains to be seen how much of the molding sand in a mold deteriorates at any one time of filling of molten steel.

The mold face is subjected to high temperatures and undoubtedly reaches temperatures equivalent to those of the molten steel. Immediately back of the mold face, the sand is subjected to comparatively high

temperatures, which vary in their magnitude with the distance from the mold-metal interface and the volume of the casting being poured.

Experimentation[1,4,5] on various molding sands and on varying sizes of mold cavities has shown that the temperature gradient curves are similar to those of Fig. 204. It will be observed from this chart that molding sand is a good insulator, and that only the sand grains very close to the mold-metal interface are altered by the heat of casting, and also that the clay bond is permanently altered at distances up to 1 or 2 in. inward from the mold-metal interface in most casting practices.

The maximum temperature curves for various-size castings are shown in Fig. 205. The heat insulating properties of the sand, such as deter-

Time Minutes.
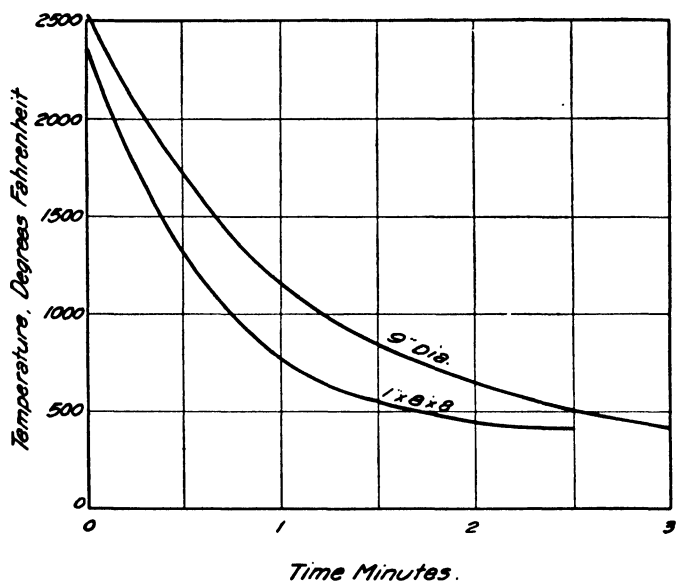
Fig. 205.—Maximum temperature reached by sand in mold for steel castings of various sizes. (*Dierker.*[1]) (*Briggs and Gezelius.*[5])

mined by the grain size, permeability, and molding material, will influence the heat flow, as illustrated in Figs. 65, 66, and 67. Thus, the maximum temperature reached in various parts of the mold will depend upon: (1) the distance from the mold-metal interface, (2) the amount of sand surrounding the castings, (3) the type of sand and amount of moisture content, and (4) the shape or volume of the casting. A complete investigation of mold temperatures would of necessity take these variables into consideration, although a study of the above charts reveals that the temperature reached by the bulk of the sand is surprisingly low.

Heat is conducted from the mold-metal interface through the mold by either of two methods: (1) convection, owing to the escape of hot air and

gases; and (2) conduction, resulting from heat transfer from one grain of sand to another. Heat conductivity in molds can be reduced by decreasing the permeability or fineness of the sand, and by selecting molding materials of low heat conductivity. Dry-sand molds transmit heat more readily than green sand.

The amount of heat flow in a mold will depend upon: (1) volume and shape of casting, (2) thickness of the mold, (3) length of time between pouring and shaking out of the mold, and (4) type and construction of the flask. Point (4) is of importance only if the mold wall is approximately 2 in. or less in thickness, or when large castings are permitted to cool in the mold.

Although the bulk of the sand in a mold does not reach very high temperatures, that portion of the mold adjacent to the mold-metal interface is subjected to very high temperatures; and since this material has considerable influence on the appearance of the casting surface, it is important that attention be given to the properties that molding sand possesses at these elevated temperatures.

Refractoriness.—The refractoriness of a molding sand is the maximum degree of heat that it will withstand without showing signs of fusion. It is important to foundrymen to know the lower limit of the fusion range, *i.e.*, the incipient fusion or the temperature at which the sand sinters and sticks to the casting, thus making cleaning difficult. The upper range, or the true fusion point as determined by cone tests, actually measures the refractoriness of the material.

Test results[6] obtained do not point to any definite relationship between the refractoriness value (as measured by the cone test) and the sintering point. They do, however, indicate that the higher the sintering point of a material, the more easily it will be removed from the casting. It has been noted[7] that an inverse relation exists between the sintering point and the burning-on effect.

Two points are recognized in standard American Foundrymen's Association laboratory determinations of the sintering of sand:[28] an A point, which is the lowest temperature at which the platinum ribbon of the sintering apparatus makes a V when lifted off the sand specimen; and a B point, or point of incipient fusion, which is the lowest temperature at which the smaller grains begin to fuse. Little attention is given to the A point by steel foundrymen. The A sintering point is much lower than the temperature to which the sand is subjected by molten steel. The A sintering point of a sand that peels perfectly from a casting may well be 800°F. lower than the pouring temperature of the metal.

A microscopic study of the sand grains[29] shows that no change occurs in the sand grains as units until temperatures much above the A sintering point are reached. The A sintering point is the temperature at which

the coating or clay substance on the sand grain has sintered or softened sufficiently to adhere to the platinum ribbon with enough force to bend it when it is lifted. The sand grains as a whole have not been changed.

At the B point definite fusion of the smaller grains takes place. At temperatures slightly over the B point, all the grains fuse as a mass. The B point is not a true melting point, for silica melts over a wide range of as much as 400°F., depending on a number of variables.

Experimental studies made by the Sintering Test Subcommittee of the American Foundrymen's Association[30] have correlated the B point with actual behavior of the sand when it comes in contact with molten steel. They noted that as the sand fuses, voids open up, and excessively fused sand has large-size voids. Voids appearing when sand fuses are caused by the increase in specific gravity of fused silica, as compared with a mass of individual sand grains. For example, a molding sand will have a specific gravity of 1.70, whereas fused silica has a specific gravity of 2.65. This increase in specific gravity means that the fused sand would occupy only 64 per cent as much space as the unfused sand; the difference, or 36 per cent, exists as voids.

Sands high in silica do not melt, as do metals, into liquid phases with viscosities approaching that of water. On the contrary, liquid silica just above the liquidus temperature is almost as viscous as when solid. Silica further differs from a metal in that melting occurs only at the outside surfaces of the particles. This means that with increasing temperature, the smaller particles of molding sand melt before the larger particles show much fusion.[30]

The behavior of the bonding materials of molding sands is an important factor in sintering, since they act as a flux. It is stated by Ardenne and Endell[31] that the cleaning costs lessen as the sintering action becomes stronger—*i.e.*, as the bonding action exercised by the fluxes on the quartz is increased. The clay- and bentonite-bonding materials at the high temperatures form a semimolten phase interspersed between the still solid and high-melting-point quartz grains.

It was concluded from the work of Ardenne and Endell that the most refractory sand is not necessarily the most suitable for foundry use, but rather those sands that, in view of a critical content of the bonding fluxes, are able to cause a cementing of the sand particles when heated by the hot metal so that the sand can be removed easily in the cleaning room. The bonds of the best steel-foundry sands melt easily and quickly at 2280°F., to form a spherical globule of molten material. Sands unsuitable for steel-foundry use are found to contain acicular masses at this temperature, adequate fusion taking place at about 2550°F.

In general, it may be said that the sintering point of a bonded sand increases with grain size, although it also increases with the addition of

silica flour (fines). Also, the sintering point for molding sands decreases with increasing amounts of natural or added bond contained.

Sintering temperatures for steel sands range from 2450 to 2650°F. for the *A* point, and from 2600 to 2750°F. for the *B* point. Sand possessing a higher sintering temperature consists of crude silica or washed silica. The maximum sintering temperature of any sand is obtained with a minimum clay content. The sintering temperature of the sand decreases rapidly as reclaimed sand is added to the facing. Bentonite-bonded washed silica sand will usually have a higher *B* point than clay-bonded sands.

The effect of additions of various materials to a washed silica sand (98.1 per cent silica) was studied by Caine.[41] Kaolin-clay and western-bentonite additions lowered the *B* sintering point from 130 to 160°F. for 5 per cent additions of the bond. The kaolin lowered the *B* sintering point the lesser. Additions of 10 per cent lime or 5 per cent dolomite must be present in the AFA clay fraction before the *B* sintering point is lowered enough to cause appreciable fusion resulting in adhering sand. Additions of as much as 10 per cent iron oxide or of 25 per cent alumina have but little effect on the *B* point.

The addition of sodium salts decreases the *B* sintering point greatly, an addition of 2 per cent borax reducing the *B* point of a silica sand by 350°F. Sands consisting of a considerable proportion of feldspar grains and showing chemical analyses with quantities of from 5 to 20 per cent of $CaO + MgO$ or K_2O and Na_2O will have *B* sintering points at from 2250 to 2500°F., which is considerably below those of sands consisting of nearly pure silica.

A report from 17 foundries[1] on casting defects caused by sand showed that the probability of producing defective castings was 33 per cent when the sintering temperature (*B* point) was 2700°F. or over, but that this value was increased in proportion to the drop in sintering temperature when it fell below this temperature.

Closely connected with the refractory property of sands is the ability of the sand to withstand the pressure of the metal and the metal fluidity and prevent it from flowing between the grains. Grains of sand having a low sintering point might easily fuse together and close the passages between them and prevent the flow of metal into the sand. In the case of high-sintering-point sands, this fusion might not take place and metal penetration would be noted. The case of metal penetration and the use of mold washes will be discussed at the end of the chapter.

Durability.—Since heat affects the sand grains and the bond of mold sands, it is expedient to know the lasting qualities of sand that are used repeatedly. This ability of using sand without additional bonding material is known as "durability." Three methods are followed in an attempt

to determine the durability of molding sands: (1) the mold-test method, (2) the oven-test method, and (3) the hydration and dehydration method.

The test that most closely approximates actual foundry conditions is to prepare and repeatedly pour small test molds, the sand being retempered and tested between the pourings with no additions except tempering water.

Test results on molding sands have led Schubert[26,27] to conclude that it is not possible to predict with accuracy the life of a molding sand by any of the three test methods generally used. The durability of a molding sand depends upon the physical and chemical properties of the minerals found in the bonding substance.

The durability of five molding sands was studied by Co Van[32] by the repeated pour test, using cast iron. Three of the sands were synthetic-bonded and two were natural-bonded sands. Each of the synthetic sands was bonded with one of the three clays: kaolinite, montmorillonite, or illite. All sands were bonded to a green compression strength of 6 p.s.i. The durability values obtained on the naturally bonded sands showed that the amount of natural-clay bond needed to produce the required 6 p.s.i. green compression strength throughout the test series was more than twice that required for synthetic sands to maintain the same strength. The amount of clay additions required was as follows: kaolinite, 122 g.; montmorillonite, 132 g.; illite, 181 g.; midwestern natural sand, 648 g.; eastern natural sand, 660 g.

Hot Permeability and Gas Evolution.—The term "hot permeability"[9] has been suggested to describe the permeability of molding sand at temperatures similar to those in the mold. A considerable difference exists between hot permeability and permeability of the same sand at room temperature. The permeability value for mold and core materials is decreased with the increase of temperature. At a temperature of about 350°F., the permeability is low, but it is usually increased at temperatures of 550°F. before falling off gradually to a low figure at 1800°F.[10] as shown in Fig. 206.

The decrease in permeability may be owing principally to the effect of gaseous expansion, and not to any pronounced structural change such as expansion or contraction of the sand.[11] The manner in which air expands with temperature is shown in Fig. 206. Thus, this chart suggests that if the molding sand is high in gas-forming materials, a high mold permeability at room temperature is essential since this permeability is somewhat related to the permeability obtained at high temperatures.

It is stated by Dietert and coworker[33] that at high temperatures of testing, such as 2500°F., there may be some relationship between the expansion of the sand and hot permeability. It appears from Fig. 207 that maximum expansion and maximum hot permeability occur at the

same time. Then, as the sand contracts, causing a reduction in expansion, the hot permeability shows a reduction.

It is believed by Dietert that direct application of hot permeability values could not be made unless information on the amount of gas gener-
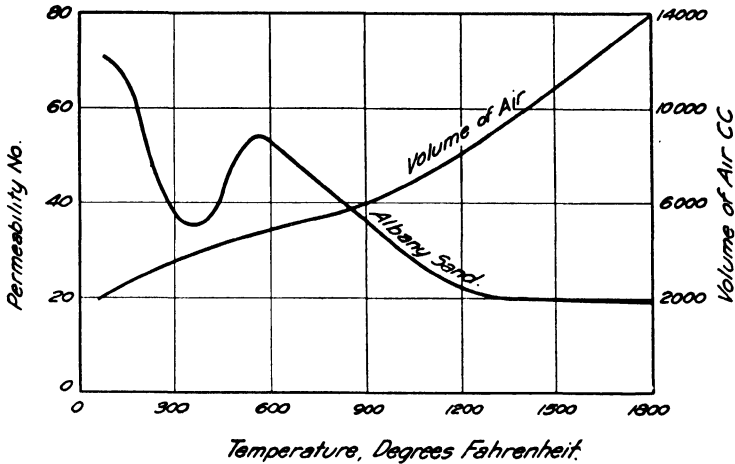
Fig. 206.—The effect of heat on the permeability of molding sand. Albany molding sand, AFA 98 fineness, clay content 10.8 per cent. (*Saunders and Saunders.*[10]) (*Hudson.*[11])
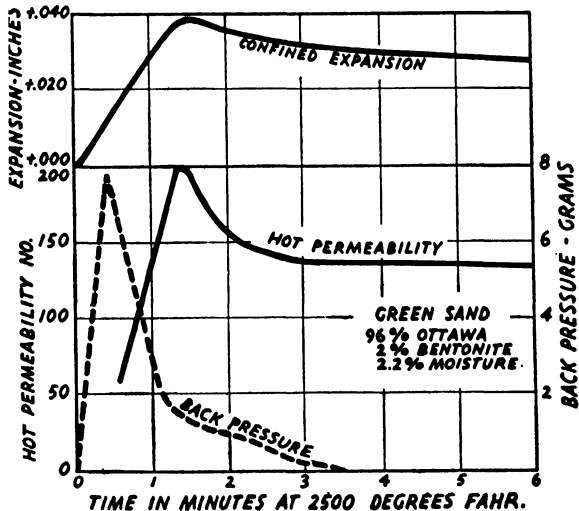
Fig. 207.—The behavior of confined expansion, hot permeability, and gas pressure on a synthetic sand at 2500°F. (*Dietert, Doelman, and Bennett.*[33])

ated were also given. For each particular volume of gas generated by the heat of the molten metal, a certain gas pressure will build up until the mold is capable of carrying off the gas. This mold capability will depend upon the hot permeability of the mold. Thus, the greater the gas volume

from moisture and combustible materials for a given hot permeability, the greater will be the gas pressure (or the back pressure).

In Fig. 207, it will be noted that the green permeability of 283 drops to a permeability of 60 in the first 30 sec. after the specimen was placed in a furnace at 2500°F. This drop is caused by a sudden generation of steam, which caused a back pressure of 8 g. As the pressure drops, the hot permeability increases and finally stabilizes as the back pressure disappears. After $3\frac{1}{2}$ min., all of the moisture and combustible material is converted to a gas. The volume of gas that is generated at the mold-metal interface and the character of the hot permeability will cause a gas

The effect of moisture contents on the amount of gas produced in synthetic green sand when molten steel is cast against it

Sand *D* contains 6% Bentonite
2% Water
Sand *E* contains 6% Bentonite
6% Water
Sand *F* contains 6% Bentonite
10% Water
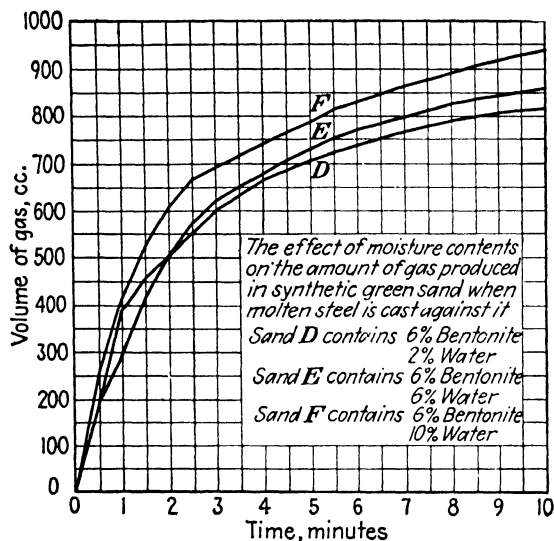
FIG. 208.—Effect of moisture contents on the amount of gas produced in synthetic green sand when molten steel is cast against it. (*Briggs and Morey.*[24])

pressure to be created that is sufficient to carry off the gas. Probably the greatest portion of the gas that is created is due to moisture in the sand.

Organic binders such as molasses, linseed oil, core oils, pitch, resin, cereal binders, and other hydrocarbon compounds form gases of carbon or methane under the action of the high temperatures of molten steel. The molten steel also turns all the available moisture into steam, with the possibility of some of it being changed to hydrogen and oxygen. The clay contents of molding sands become dehydrated under the action of heat, and the water will be converted into steam. The gases produced may, in cases of low mold permeability, result in the formation of gas cavities in the casting. Thus a knowledge of the amount of gas produced in molds is desirable.

The effect of moisture contents upon the amount of gas produced in synthetic green sand when molten steel is cast against it[24] is shown in

Fig. 208. From this graph it may be seen that the volume of gas increases slightly with increase in water contents. An increase in bentonite-binder content does not increase the gas volume. Figure 209, however, shows that small amounts of organic binders cause a tremendous increase in the amount of gas produced.[24]

THE EFFECT OF ORGANIC BINDERS ON
THE CUMULATIVE VOLUMES OF GAS
PRODUCED IN SAND WHEN STEEL IS
CAST AGAINST IT.

SAND G CONTAINS 4% BENTONITE
4% WATER
0% "BINDEX"

SAND H CONTAINS 4% BENTONITE
4% WATER
1% "BINDEX"

SAND I CONTAINS 4% BENTONITE
4% WATER
2% "BINDEX"
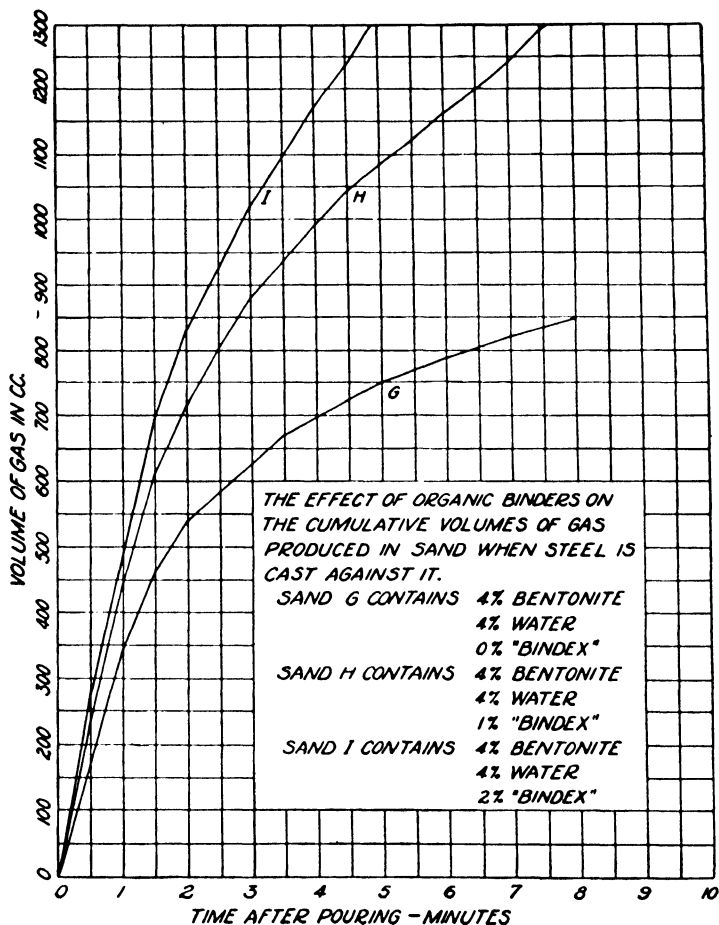
FIG. 209.—Effect of organic binders on the cumulative volumes of gas produced in green sand when steel is cast against it. (*Briggs and Morey.*[24])

In Table LXIII are presented sands of different mixes with sufficient tempering water to establish a proper working condition of the sands. It will be noted that the amount of gas is related to the amount of moisture. Also, the addition of but 1 per cent of organic binder increases the gas content considerably and reduces the hot permeability, which in turn results in a greater gas back pressure.

TABLE LXIII.—GAS EVOLUTION AND HOT PERMEABILITY
Dunbeck[34]

| Type | Per cent moisture | Steam at 212°F. | Cu. ft. gas at 1800°F. per cu. ft. of sand | Permeability | | Total gas 2500°F., 4 min., cc. per g. sand | Back pressure, g. |
				1¼-in. specimen, 70°F.	1⅝-in. specimen, 2500°F., 4 min.		
5 per cent western bentonite.........	2.5	40.0	232.3	290	150	242	6.6
4 per cent southern bentonite........	2.5	41.5	247.8	294	158	258	5.8
11 per cent Ohio fire clay...........	3.5	56.5	324.8	148	111	338	12.5
5 per cent western bentonite, 1 per cent cereal......................	3.4	56.5	370.9	170	86	341	15.5
5 per cent western bentonite, 1 per cent resin.......................	3.4	56.5	351.5	140	93	366	18.0
5 per cent western bentonite, 1 per cent dextrin.....................	3.5	58.0	374.4	298	123	390	8.4
Albany (natural bonded)...........	4.8	78.0	502.4	2	0	410	400+
Ohio (natural bonded).............	7.8	124.8	778.3	40	0	812	400+

The gas content as produced in rapidly heated molds can be reduced by employing: (1) a synthetic sand; (2) bentonite bond; (3) low moisture content; (4) low, or no contents of organic binders.

Expansion and Contraction.—It has been shown previously that sand grains expand upon heating, and that clay and cereal binders contract upon heating. Likewise, it has been reported[8,11,12,13,14,35] that molding sands subject to considerable expansion and contraction are responsible for the formation of casting defects such as scabs, buckling, and rat tails. It has further been suggested[12,13,14] that by the addition of certain binders to sand it is possible to alter the expansion and contraction characteristics at high temperatures by varying the type and amount of bonding materials in order to bring the expansion and contraction to a minimum figure.

A recent set of studies on expansion and contraction of molding sands[15] gives results that conflict somewhat with previously published results. Tests were made on the free expansion and contraction of molding sands containing varying amounts of clay bond and moisture. An example of the type of curve obtained is shown in Fig. 210. It is interesting to note that in all the expansion curves studied, the contraction on cooling is less than the total expansion. Studies were made on a washed silica sand wherein the clay content, both bentonite and fire clay, was varied from 4 to 10.7 per cent and the moisture content varied from 1.9 to 4.3 per cent. The expansion curves were so nearly alike that they could not be plotted together, and all resemble Fig. 210. Thus the total free contraction is not materially changed by the amount or type of clay,

the moisture, or the degree of ramming. It was also observed that the contraction in the case of fire-clay-bonded sands was relatively small as
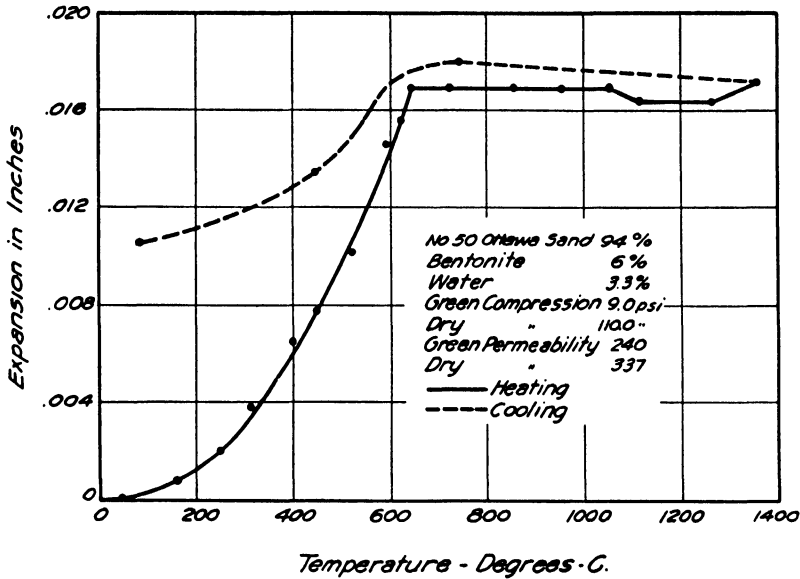
FIG. 210.—Free expansion curve of $1\frac{1}{8}$-in. normally rammed specimen. (*Ehrhart.*[15])
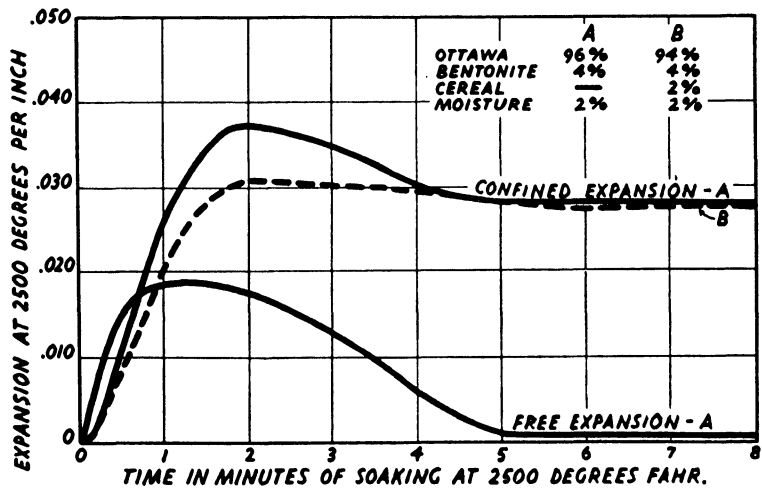
FIG. 211.—Confined expansion for green synthetic sand at 2500°F. (*Dietert, Doelman, and Bennett.*[35])

compared with that of the bentonite-bonded sands. The contraction of synthetic molding sands in general varies more than the expansion.

Most information on the expansion and contraction of molding sands has been obtained by testing sand specimens under partial restraint.

Such testing apparently gives results that are considerably different from free expansion and contraction measurements, although even in this type of testing the greatest degree of expansion takes place between 500°C. (932°F.) and 700°C. (1292°F.).[11]

Recently a confined-expansion test was developed to facilitate the practical application of expansion-test data.[35]   In the confined-expansion test, the 1⅛- by 2-in. test specimen is rammed in a quartz tube and tested in the tube.   In the free-expansion test, the sand specimen is free to expand in circumference as well as in length.   A comparison of confined and free shrinkage on the same sand is shown in Fig. 211.   The effect of an addition of cereal bond is also shown.   The confined expansion is twice that observed for free expansion.

Both Dietert[13,14] and Hudson[11,12] have been most consistent in pointing out that by the use of organic binders and by controlling the amount and type of clay binders, it is possible to arrive at molding sands with very low expansion and contraction characteristics.   Dietert and coworkers[13,35] point out that the partial restraint and confined expansion of molding sand may be reduced by

1. Increasing grain size
2. Reducing the fines
3. Addition of combustible materials (core oils and cereal binders)
4. Reducing the mold hardness
5. Increasing permeability
6. Reducing moisture of green sands
7. Additions of silica flour
8. Reducing the flowability

It has been stated[8] that the expansion and contraction of sands are decreased as the content of the clay is decreased.   It was also noted that dry sands increased confined expansion over the same sand in the green state.

Regarding the effect of moisture on the expansion and contraction of sands tested under partial restraint, Dietert, Woodliff, and Schuch[8] state that molding sands that are dried or have no free water have greater dimensional stability.   They show less expansion and contraction upon heating.   This smaller volume change will produce a mold surface where the least possibility exists of producing scabs and buckles.   They further state that the expansion of molding sand is reduced upon continued use, in that when molding sand is heated above 2400°F. it will expand only approximately one-half as much as the original sand.   Contrary results reported by Ehrhart[15] on the free expansion and contraction tests of molding sand show no differences between used and new sand.

According to Dietert[14] one of the best ways to ascertain if molding sand is subject to producing buckling or scab defects on castings is to
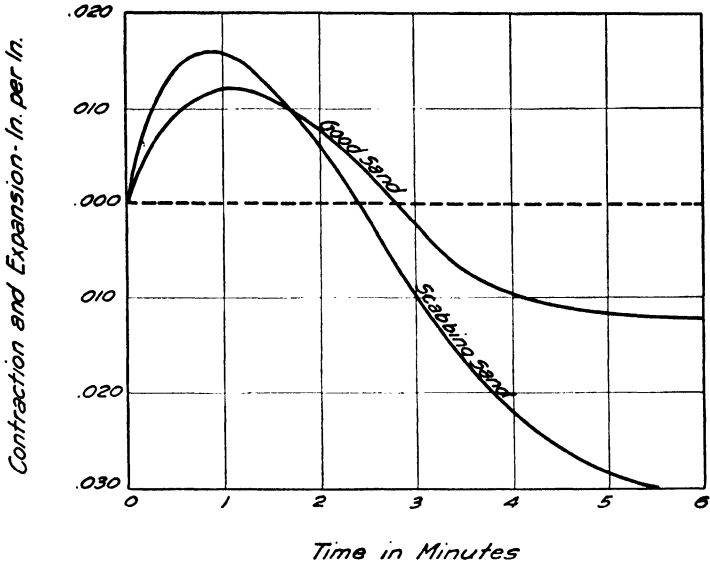
FIG. 212 — Relationship between linear change in length and time at 2500°F. (*Dietert.*[14])
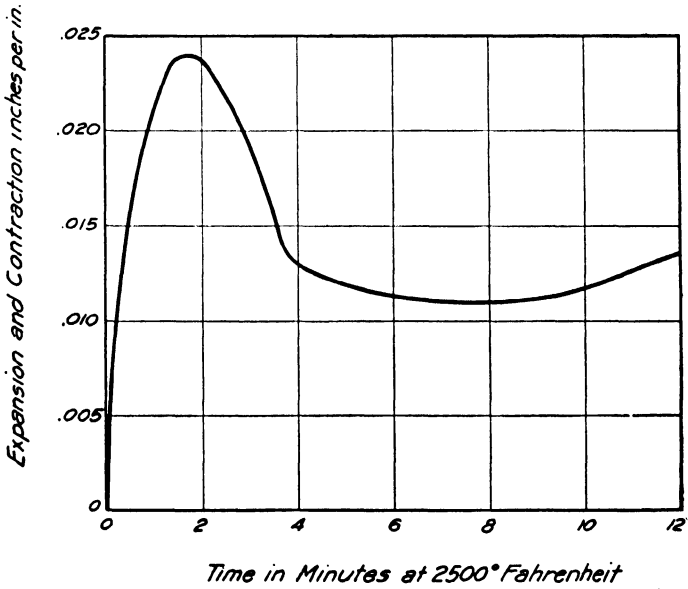
FIG. 213.—Expansion and contraction of a steel-facing sand at 2500°F.

subject the sand sample to a temperature of 2500°F. and measure the expansion and contraction that take place over a period of 5 or 6 min. Expansion and contraction relationships for a good sand[14] and one that produces scabs are shown in Fig. 212.

An expansion-and-contraction curve obtained by one steel foundry on their regular facing sand is shown in Fig. 213. The sand is of AFA 50 fineness and is synthetically bonded. Such a sand-expansion condition is comparable to the results shown by the good sand curve in Fig. 212, except that contraction conditions did not extend below the original length of the specimens.
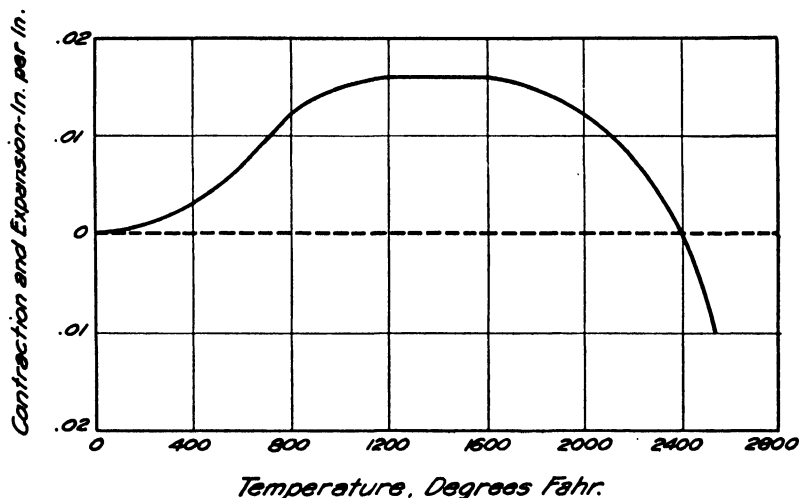
FIG. 214.—Expansion and contraction of a steel-foundry sand upon gradual heating. (*Dietert.*[14])

A relationship appears to exist between sintering point and linear change in contraction after expansion at 2500°F., for crude and washed silica sands.[14] The sintering point of the sand rises as the contraction lowers. The manner in which a steel-foundry sand expands and contracts when tested under partial restraint[14] is shown in Fig. 214.

Buckling and Scabs.—It has been stated that buckling and scab defects on castings are caused by excessive expansion and contraction of sand at the mold-metal interface.

Buckling, or "rat tail," usually occurs in lightweight castings where the temperature of the sand does not reach the contraction point, but where sufficient heat is present to expand the sand. The expanded sand must find space or the mold face will be broken.[8]

Scabs are caused by high contraction values. The high contraction of the sand causes cracks to open in the face of the mold into which molten metal penetrates. The backing sand, being at a lower temperature,

expands and tends to push the facing sand into the molten metal. This combination of excessive expansion and contraction is, according to Hudson[12] and Dietert,[13] the direct cause of the scab defect. It is their claim that these defects can be eliminated by the addition of clays or organic binders to the sand mix.

Contrary results have been presented by Buchanan[16] in that his tests show that expansion of silica sand cannot be masked by additions of other materials or by ramming. It is his belief that the amount of expansion recorded is of no consequence, and that it does not cause scabs; rather, the scabs are caused by steam from the moisture content of the sand and lack of permeability of the sands. The steam pressure is responsible for pushing the facing sand into the metal. This explanation seems to be somewhat out of line with practical results obtained by foundries in preparing proper molding sands based on expansion and contraction studies.

Summarizing the effects of temperature on the expansion and contraction of molding sands, it appears that results of testing under partial restraint and free testing are somewhat at variance. Practical considerations may more likely favor partial-restraint testing, since molding sands in a mold are seldom able to move freely. Conclusive evidence is available that, by the use of cereal binders and modifications in clay types, contraction characteristics can be modified considerably, even though expansion values cannot be changed appreciably. Likewise, until there is more evidence to the contrary, scabs and buckles appear to be caused by excessive expansion and contraction of sands at the mold-metal interface.

Effect of Temperature on the Strength of Molding Sands.—All molding sands are apparently stronger at some elevated temperature than they are at room temperature before being heated. Usually the peak in the increased strength occurs in the neighborhood of 2000°F. These high-strength sands near the face of the castings are instrumental in hindering the casting from its normal contraction. Hindered contraction of a casting results in the setting up of stresses within the casting members, and high hindered-contraction stresses are conducive to the formation of hot tears and casting warpage.

The compression strength of molding sands at elevated temperatures depends upon a number of things and, since research on this subject has been going forward only within the past few years, there are still many things that are not understood concerning the effect of variables of molding-sand composition.

Besides the strength or compression test, a heat-shock test has been devised. This test subjects the molding sand to high temperatures by placing the room-temperature specimen in a furnace that is maintained at the desired temperature. Values obtained on a molding sand tested by heat-shock[17] test are shown in Fig. 215. The compression-strength

characteristics at elevated temperatures are also shown in Fig. 215. These curves are typical of a sand mixture not affected by temperature shock. The curves at the left show the effect of the application of the heat shock. The compression strength increases until the specimen is completely up to temperature. Absence of shock effect is shown by identical strengths upon heating slowly as compared with sudden exposure to the test temperature. The curves at the right show the strength attained by the sand specimens after exposure at a certain temperature for various time intervals. The bond partially fuses and is plastic between
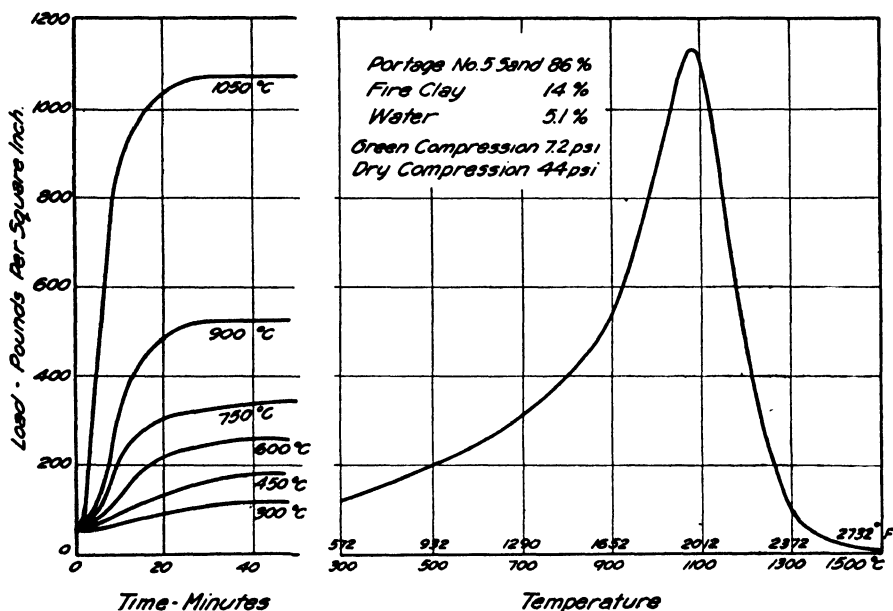
FIG. 215.—Strength values obtained by testing a sand mixture not affected by heat shock. Heat shock tests on left. Slow heating tests on right. (*York.*[17])

1075 and 1300°C. (1967 and 2372°F.). The bond acquires a glassy appearance at 1300°C.

A sand that did not withstand the shock test[17] at 600°C. (1112°F.) is shown in Fig. 216. At this temperature and temperature above 600°C., the specimen cracks when subjected to heat-shock tests. The effect of cracking is to reduce the strength of the specimen. The true strength of the sand mixture can be obtained only by slow heating, as shown by the curve to the right in Fig. 216.

Test results have shown that some sand mixtures are sensitive to heat shock, while other mixtures are unaffected and develop the same strength under shock-test conditions as they did when heated slowly. Neither the bentonite- nor the fire-clay-bonded sands as a group are unaffected by heat shock.

Molding sands at the mold-metal interface that fail because of heat shock may be responsible for mold spalling and veining (metal penetrating cracks in the mold), thus producing rough and poorly appearing casting surfaces. Commercially successful foundry sands do not necessarily show absolute freedom from cracks on shock heating. However, it has been found[19] that those sands that do not crack on shock heating produce more satisfactory casting-surface finishes.

Moisture influences the hot strength of molding sand up to temperatures of approximately 2000°F., the strength increasing as the moisture content increases. The effect of moisture is eliminated when a temperature of 2500°F. is reached.[18]
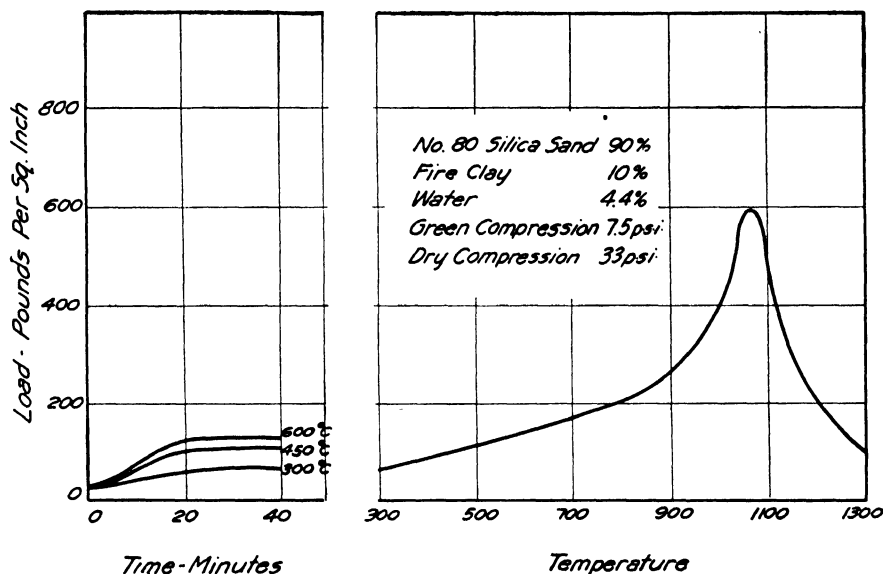
Fig. 216.—Sand mixture that did not withstand shock tests at 600°C. (1112°F.) and over. The true strength is shown by the slow-heating curve at the right. (*York.*[17])

Dietert and Woodliff claim[18] that, for all molding sands, the average hot strength decreases as the grain size increases. Thus, to increase the collapsibility of a molding sand, the grain should increase in coarseness. This conclusion may or may not be a fact, since during testing other variables such as moisture and bond content, as well as grain size, were also varied. Hot strength of a sand also increases at a faster rate than the rate of increase of the green or dry strength, while increasing mold hardness also increases the hot strength.[18]

Mixtures of silica sand and bentonite invariably disintegrate on cooling to room temperature after hot-strength testing, and silica sand-fire-clay mixtures are very fragile after cooling. Natural-bonded sands generally have considerable retained strength after cooling.[19]

The hot strengths of steel-molding sands will vary considerably as is shown in Fig. 217. Molds that must withstand a considerable amount of metal weight should have fairly high hot strengths, whereas light, complicated castings should be produced in sands of low hot strength, in order to prevent the formation of high mold restraints against casting contraction. These curves give an idea of the amount of mold resistance that these sands set up and they therefore show the possibility of hot-tear-formation characteristics.
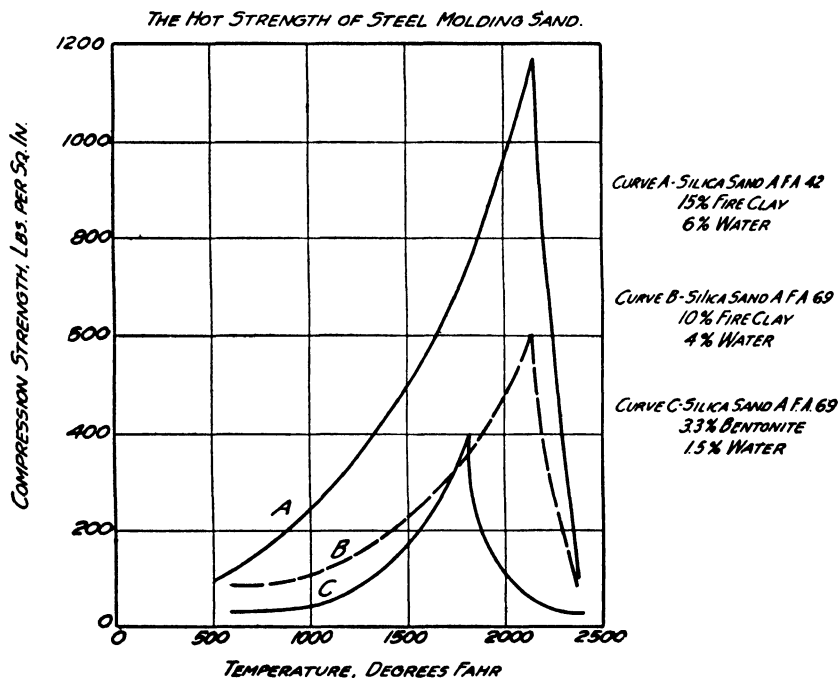
FIG. 217.— The hot strength of steel-molding sands. (*York.*[17])

The effects of various binders on the hot strength of silica sand[20] of a uniform AFA 53 fineness are shown in Fig. 218. The addition of a cereal binder appears to lower the hot strength. The room-temperature properties of the synthetic-bonded sands illustrated in Figs. 217 and 218 are shown in Table LXIV.

The effect of exposing the sand for various periods of time at any testing temperature[36] is illustrated by Fig. 219. Testing was carried on at 2000°F., and different bonding materials were used for an AFA 60 washed silica sand at a moisture content of 5.3 per cent. The chart shows that a green-sand test specimen inserted into a furnace at 2000°F. will increase in strength as the heat penetrates the specimen and brings it entirely up to a temperature of 2000°F. Strength changes in the last 6 min. of a 20-

min. holding time are not marked.   Figure 219 also shows the effect that different types of bond have on hot compression strength.
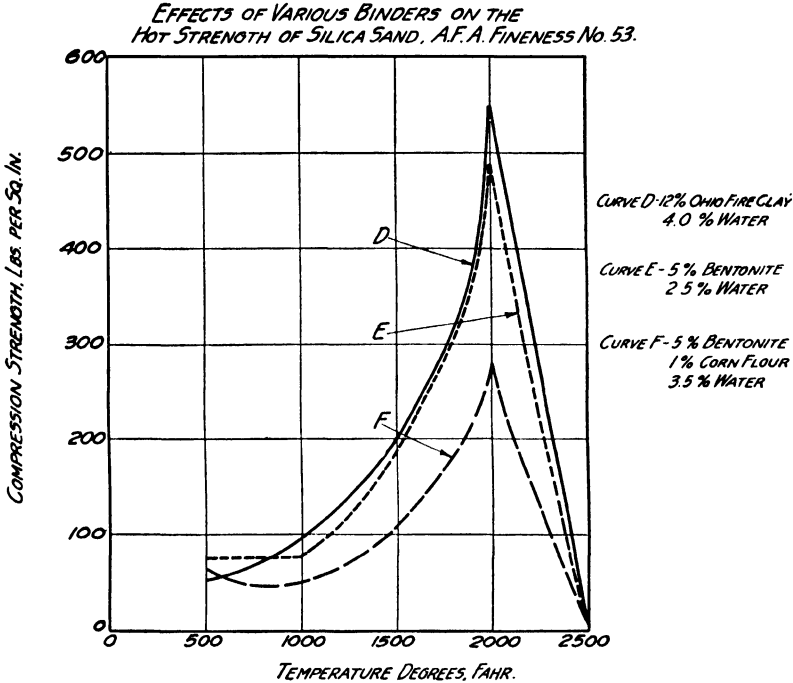
FIG. 218.—Effects of various binders on the hot strength of silica sand.   (*Dunbeck*.[20])

In tests carried on for different soaking or exposure times at any one temperature, there may be found secondary peaks in the curve, especially at time intervals of 1 and 2 sec.   These are undoubtedly caused by rapid

TABLE LXIV.—PROPERTIES OF VARIOUS SYNTHETIC-BONDED SANDS

	A	B	C	D	E	F
	Fire clay	Fire clay	Benton-ite	Ohio fire clay	Benton-ite	Bentonite and corn flour
Bond, per cent.............	15	10	3.3	12	5	5
Cereal binder, per cent.....	0	0	0	0	0	1
Moisture.................	6	4.4	1.5	4.0	2.5	3.5
Permeability..............	205	67	112	85	167	152
Green compression..........	7.3	7.5	7.6	8.5	9.6	9.3
Green deformation..........	.....	....	.....	0.019	0.012	0.027
Dry compression...........	61	33	9	71.5	83.0	87
AFA fineness No...........	42	69	69	53	53	53

drying of a green sand, thereby producing a dry strength. Secondary peaks are also produced whenever silica flour or fine silicon material is present in appreciable amount.[35] Soaking time at temperature affects dry sands differently from green sands (see Fig. 220). Dry-sand molds have considerable dry compression strength at room temperature. This strength is reduced in the first few minutes of heating. On continued heating, an increasing strength results.

Observations by Dietert and coworkers[35] reveal that for short soaking periods, the hot strength of bentonite-bonded sand with a limited silica-
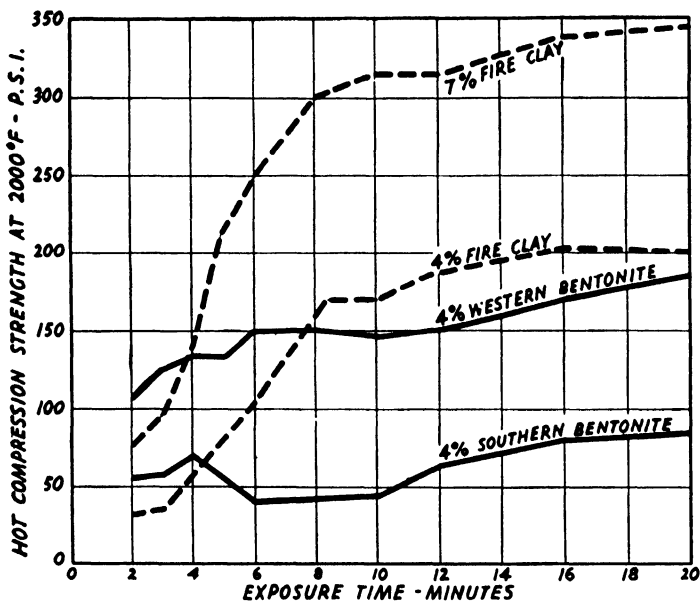
Fig. 219.—Effect of additions to New Jersey silica sand No. 60 and 5.3 per cent moisture content on the hot compression strength at 2000°F. Green-sand specimen tested. (*Williams.*[35])

flour content is of the decreasing order for dry sands, while of an increasing order for green sands. Under long soaking periods such as 12 min., the difference between the hot strength of a green sand and dry western-bentonite-bonded sand is not great, as illustrated in Fig. 220, when the temperature is above 1000°F.

It has been stated by Young[37] that the degree of ramming of the sand (ramming intensity) has an effect on the hot strength of the sand. An increase in ramming intensity seems to lower the temperature at which is recorded the peak hot strength. He also pointed out that hot strength varies with moisture and percentage of bond in much the same manner as dry compression varies with moisture and percentage of bond.

Attention is directed to a new trend in the bonding of sands, the use of

combinations of clays instead of a single clay. Figure 221, taken from the results of Dunbeck,[20] shows the gradual increase in hot strength from southern bentonite, through combinations to western bentonite. Low hot-strength southern bentonite may be responsible for mold cutting and washing, whereas high hot-strength western bentonite may produce castings with hot tears. A combination of the two may be made to yield the exact strength required. Figure 222 shows that combinations of kaolinite clay and western bentonite can be used to produce hot strengths
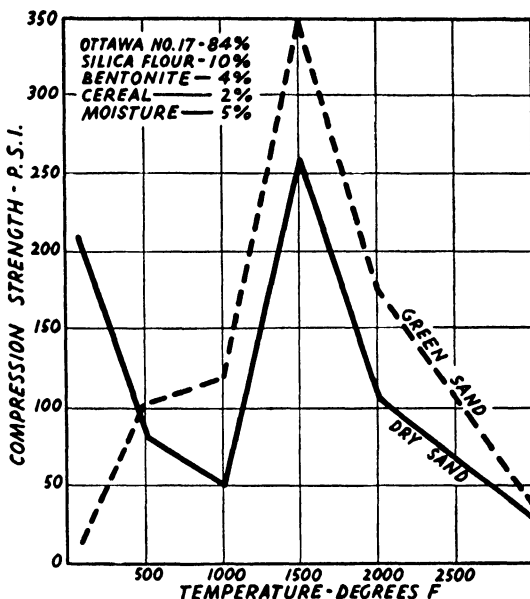
FIG. 220.—Comparison of hot strength of green and dry sand at 12-min. soaking time. (*Dietert, Doelman, and Bennett.*[35])

that give a peak strength about twice as high as either component. Such combinations should have excellent applications for heavy steel castings.

Silica flour is added to molding sand to increase the refractoriness of the sand and to stop metal penetration. By adding silica flour, the other properties of the sand are changed and the following values increased: moisture, green compression, contraction, toughness, dry compression, density, hot deformation, and hot strength. Permeability and flowability decrease as the silica-flour content increases.[21] The manner in which silica flour increases the hot strength is shown in Fig. 223. A 10 per cent addition of silica flour causes the hot strength at 2500°F. to increase from 7 lb. for no silica flour to 40 lb. This value is so high that it may cause hot tears in castings. The 20 per cent addition raises the value to 170 lb. at 2500°F. This value is entirely too high for a good molding sand.

When the hot strength at 2500°F. is below 3 p.s.i., the mold surface is subject to erosion; if the hot strength is more than 15 p.s.i., the mold surface will withstand a large quantity of metal before cutting takes
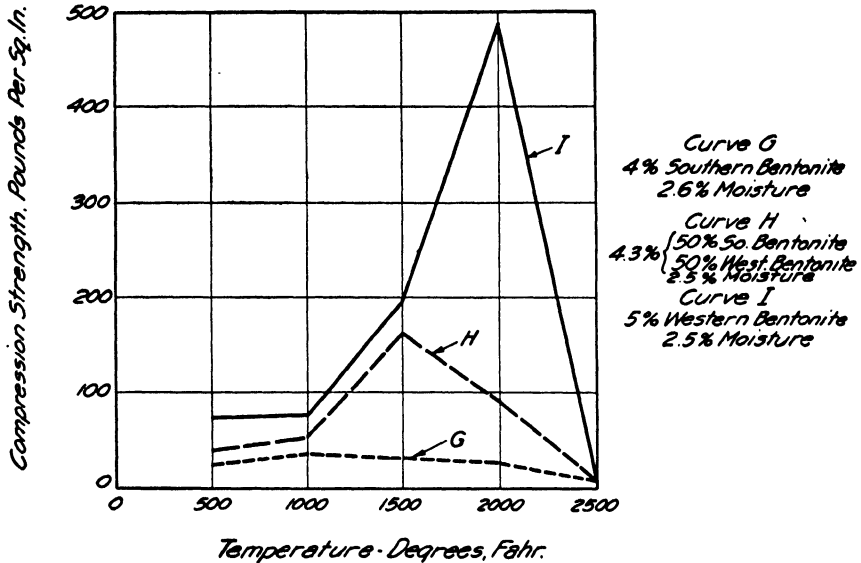
FIG. 221.—Hot strength of combinations of western and southern bentonite clays. (Dunbeck.[20])
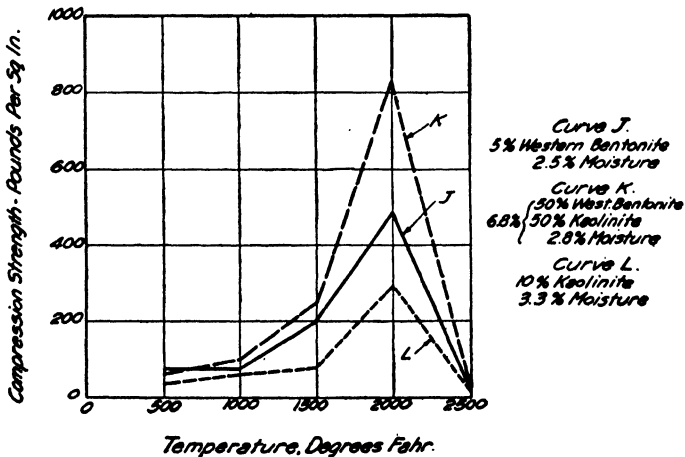
FIG. 222.—Hot strength of combination of kaolinite and western bentonite clays. (Dunbeck.[20])

place. Large steel castings probably should be produced in sand having a hot strength of 10 to 15 p.s.i.[21] The maximum hot strength of sands for steel castings probably should not be over 25 p.s.i.

A study[38] was made to determine the accuracy of compression testing at a temperature of 2500°F. A single laboratory was able to check its results within 15 per cent. Using the same testing technique as for the single laboratory test, four other laboratories testing identical specimens obtained erratic values, percentage of deviation from the average being as high as 40. This deviation was largely caused by the great temperature gradients in the testing furnaces. Gradients of 42 to 87°F. were found within a single furnace, and a maximum difference of the average temperature from furnace to furnace of approximately 100°F. Such
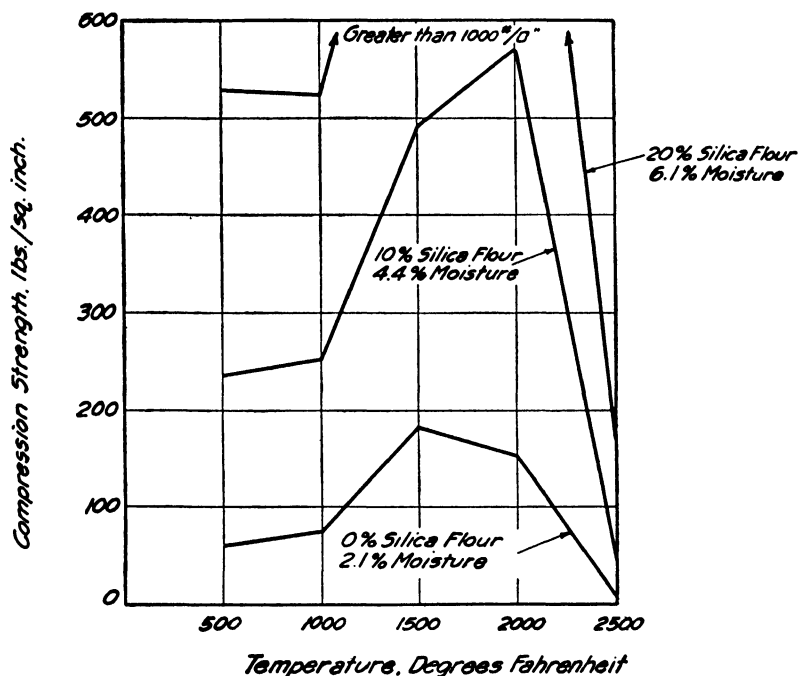
Fig. 223.—The effect of silica flour on sand bonded with 4.5 per cent western bentonite. (*Dietert and Curtis.*[21])

results were characteristic of all furnaces in the industry prior to October, 1944, when the Dietert Company, manufacturers of the furnaces, advised of methods to correct this condition.

Core Sands at Elevated Temperatures.—A core that is surrounded by molten steel usually expands at first, and then it collapses after the bond is destroyed by the heat. The manner in which cores collapse can be studied by the heat-shock test, wherein the core sample is subjected to a heat shock of 2500°F. or any other desired temperature, and the time for it to collapse is noted. A load of 4 oz. per sq. in. is applied to the core at 2500°F., and the time in minutes required to collapse the core specimen fully or to cause a contraction of 0.070 in. is recorded as the
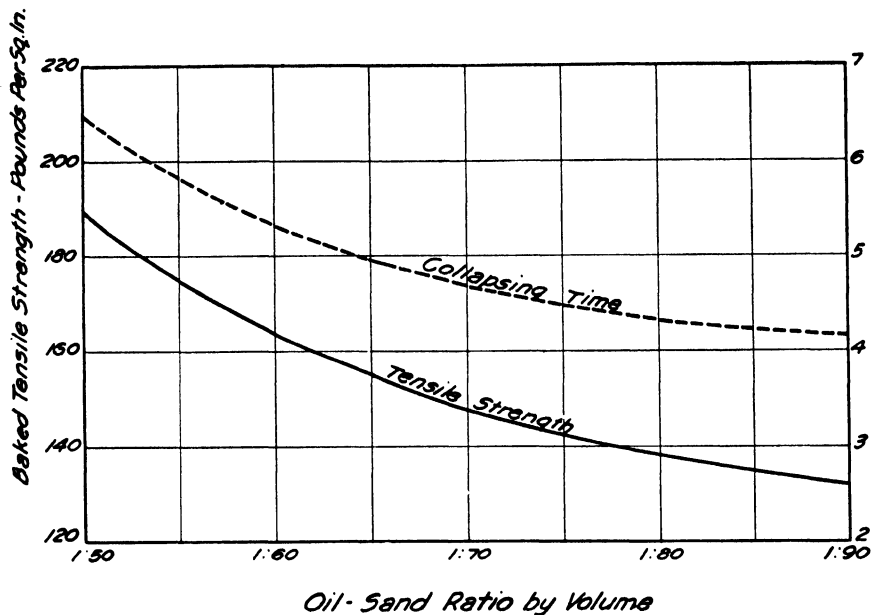
FIG. 224.—Relation between core collapsibility and tensile strength of baked cores. (*Dietert.*[22])
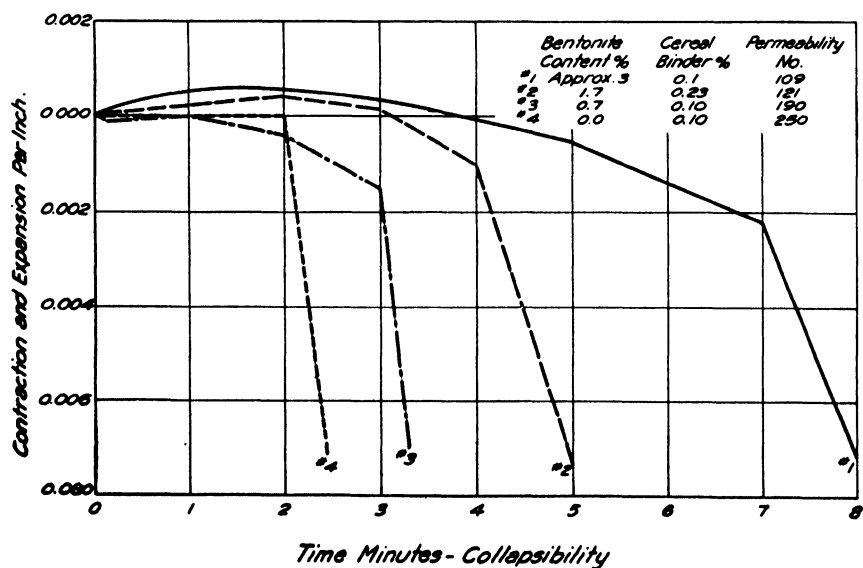
FIG. 225.—Volume change and collapsibility of baked cores. (*Dietert.*[22])

collapsing time. The collapsing time of cores usually increases as the room-temperature tensile strength of the baked core increases.[22] This relationship is shown in Fig. 224. As the tensile strength of a core decreases, the expansion generally decreases rapidly.

The rate at which a core collapses should be controlled, if best casting conditions are to be attained. If a core collapses too rapidly, there is a possibility that cuts and washes will be present on the core face, and eroded sand may be included in the casting. Should the core collapse too slowly, the casting may exhibit hot tears owing to stresses set up by high core resistance to the casting contraction.

The rate at which a core collapses as it is heated by molten metal may be varied not only by controlling the strength of the core but also by controlling the permeability and clay content of the core mixture.[22] An example of the variations in collapsibility is shown in Fig. 225. It may

TABLE LXV.—RATE OF COLLAPSIBILITY FOR CORE OILS AND AMOUNT OF CORE GAS
WHEN BAKED AT BEST BAKING TIME AND TEMPERATURE
Oil added, 0.76 per cent

Type	Collapsing time			Core gas, cu. cm. per g.
	1500°F.	2000°F.	2500°F.	1800°F.
1. Raw linseed	2′ 15″	1′ 20″	1′ 18″	8.26
2. Boiled linseed	2 27	1 58	1 42	7.93
3. Linseed + fish + light mineral oil + resin				
Oil a	2 20	1 34	1 27	7.19
b	2 17	1 33	1 27	7.19
c	2 3	1 10	1 23	7.30
d	1 55	1 17	1 18	7.44
4. Linseed + light mineral oil + resin	2 22	1 36	1 22	7.63
5. Linseed + fish oil + solvent	2 4	1 0	1 2	6.13
6. Linseed oil + mineral polymers				
Oil a	2 30	1 32	1 26	9.00
b	2 23	1 44	1 35	8.88
c	2 44	1 50	1 37	9.20
7. Mineral polymer type				
Oil a	2 15	1 36	1 25	7.14
b	2 20	1 45	1 36	6.38
8. Mineral polymers + vegetable oil + resin				
Oil a	2 6	1 12	1 3	9.36
b	2 5	1 24	1 25	7.63
9. Light mineral oil + resin	2 3	1 9	51	7.49

* Dark-colored core oils.

be noted that core No. 1 contained considerable clay and a permeability of 109, requiring 8 min. to collapse, whereas core No. 4, with no clay bond and 250 permeability, collapsed in 2.5 min.

Core oils do not vary greatly in their effect on core collapsibility at elevated temperatures. As may be seen from Table LXV, there is no indication that one type of core oil excels in the rate of collapsibility. Core oils that have a relatively fast breakdown time at 1500°F. do not necessarily collapse rapidly at 2500°F. It may be concluded from a study of the oils listed in Table LXV that oils of different compositions have close average collapsibilities at elevated temperatures.

The collapsibility time of cores can be decreased by

1. Use of proper clay or combination of clays
2. Reduction of clay content
3. Reduction of moisture content
4. Reduction of mold hardness
5. Reduction of pitch content

The hot strength of cores may be decreased by the addition of cereal binders to the core mixture. The addition of iron oxide to core mixtures increases their hot strength,[23] as shown in Table LXVI.

TABLE LXVI.—EFFECT OF IRON OXIDE ADDITIONS ON THE HOT STRENGTH OF CORES

Iron oxide additions, per cent	Baked-core tensile strength, p.s.i.	Hot strength at 2500°F., lb.
0	120	4
1	116	17
2	102	32
3	93	71
4	85	112
5	77	200

Regan[23] states that controlling core mixtures entirely by the hot-strength test has proved advantageous in practice. The fact that hot strengths could be controlled solely by the addition of iron oxide is of considerable value, in that for different core requirements it is not necessary to change the basic core mixture but merely to make varying additions of iron oxide. Iron oxide additions to cores prevent metal penetration on the inside of castings, where sharp corners or abrupt changes occur.

Core gas that is liberated during the pouring of a casting should be held to a minimum since the burning of the gas produces an undesirable smoke. The amount of core gas generated can be reduced by increasing

the grain size of the sand. The permeability of the core increases in proportion to coarseness of the sand. The greater the permeability, the better the core will be baked and the oil oxidized. This, in turn, means that less gas will be produced when the casting is poured. The quantity of gas released apparently increases in a straight-line function, according to Dietert,[39] as the fineness of the sand is increased. Too coarse a sand may not be used, or the rate of heat penetration will be high and core collapsibility may be earlier than is desired.

The quantity of gas generated by a core depends upon the amount of core oil added. For example, the gas content of a core of a sand-oil ratio of 98 to 1 by weight is found to be 7 cu. cm. per g., whereas when the ratio is 40 to 1 by weight, it is 17 cu. cm. per g.

The gas content of a core may be reduced by baking the core at a temperature above that at which it develops its maximum core strength. If this is done, some sacrifice in the baked strength of the core is made; but the gas generated will be nearly a third less. The rate at which the gas in a core is liberated may be as important as the total volume of gas liberated. The rate of liberation of gas will depend upon the composition of the core. For example, oil binders have a faster gas liberation than has pitch.

The core-gas content of various core oils, when baked at their best baking cycle for time and temperature, does not vary greatly, as may be observed in Table LXV. There is an indication that the dark-colored core oils do not bake out as thoroughly as do the light-colored core oils, since the dark oils show an average of about 10 per cent greater core gas. Cereal binders also are responsible for the evolution of considerable amounts of core gas.

CORE AND MOLD WASHES

Increasing requirements by consumers of steel castings in regard to casting surfaces have made it imperative that the steel foundries strive persistently to produce cleaner and smoother castings, free from surface blemishes. One of the most important means toward accomplishing improved casting surfaces is the proper application and use of core and mold washes.

Washes in general rely upon the use of refractory material having a fusion point at least as high as, and in most cases higher than, the facing sand of either molds or cores. The refractoriness of the wash, together with its fineness characteristics, is responsible for a minimum of metal penetration into the molding material and allowing the sand to peel freely from the casting, leaving a surface that is smooth and readily adaptable to machining requirements.

The requisites for a wash suitable for use with steel are that it must
1. Form an impermeable skin to resist metal penetration
2. Have a low surface friction
3. Be refractory
4. Form a strong coating firmly attached to the mold or core
5. Contain as little gas-producing ingredients as possible

The direct functions of core and mold washes have been listed by steel foundrymen and are reported to be
1. To prevent metal penetration
2. To eliminate cutting
3. To prevent scabbing
4. To give smooth casting appearance
5. To obtain clean casting surfaces
6. To modify mold atmosphere
7. To assist in the peeling of sand from the casting

Methods and Their Application.—Of the currently used washes, those with silica as their base are the most popular.   Silica flour of a fineness of −200 mesh usually constitutes the base material, and this is mixed with a binder and thinned with water to the proper consistency.   There are many mixtures of mold and core washes.   A number of foundries prepare their own mixtures.   Some purchase a completely compounded wash; and others purchase a prepared paste and mix silica flour and water into it, in accordance with their own requirements.   The proprietary washes are of three types: semiplastic, powder, and liquid.   Practically all of the proprietary washes are ready for use after being mixed with water and brought to the proper consistency.

Foundries that prepare their own mixtures start out with silica flour and generally add bentonite as the main clay binder.   Other additions such as cereal binders, pitch binders, molasses, linseed oil, core oils, and the like are made, depending on the particular foundry needs.   A mold and core wash consisting of powdered bakelite and alcohol is used in some foundries; others may use a wash consisting of alcohol and Truline binder. In most cases the same wash is used for both molds and cores.   Sometimes this wash is thinned and further additions of silica flour are made. Again, because of the size of the core or its position in the mold, an entirely different wash is used.

A questionnaire on the reasons for preferring a proprietary wash to a foundry-mix wash revealed that over a period of use a more uniform wash —and thus more uniform results—can be expected from the proprietary washes.   Another feature in favor of the proprietary washes is that they require less attention and do not settle so readily as the other washes.

Advocates of foundry-prepared washes believe that most purchased washes do not penetrate the sand, that they peel off and cause scabs, that

they are difficult to mix, that they have too much gaseous material included, and that the foundry-prepared washes make a better mold-metal interface.

Just what are the ingredients of proprietary mold washes, other than silica flour, is difficult to say without complete analysis, since the manufacturers of these washes are rather diffident in reporting compositions. Some use emulsified oils; others include resins; and still others include organic binders such as molasses, flours, wood pulp, and the like.

Foundry-prepared mixtures usually contain a clay type of binder that varies from 1 to 5 per cent. Foundries in one district reported that about 3 per cent bentonite is their average binder content. These same foundries are about equally divided in opinion as to the value of core oils in washes. Those using core oils in washes are of the opinion that only small percentages should be used in the mix.

It has been stated that molasses and core oils are sometimes used as binders for core washers. These constituents are not believed to be quite so satisfactory as bentonite binder, particularly in thick castings where a large area of the mold is exposed to the rising metal. The molasses or oil binders burn out and allow the silica flour to fall in flakes or dust, to form dirty areas or scabs on the casting. These materials are detrimental also because they are gas-forming.

Mold and core wash is applied in steel foundries throughout the country by the following methods: dipping, swabbing, brushing, or spraying. Spraying and swabbing are the most commonly used methods. The degree of viscosity of the wash used for these four methods of application, based on submitted results from foundries, is as follows:

	° Bé.	Average
1. Swabbing	40.70	50
2. Brushing	40.70	50
3. Dipping	44–50	47
4. Spraying	40–47	42

In general, most foundries attempt to maintain a standard viscosity of washes, whether they be prepared from the foundry mix or made up from purchased washes.

It is good practice to make preliminary tests on washes, to ascertain whether or not they stand up under conditions desired. These tests usually consist of applying the wash to sample molds or cores and observing visually the manner in which molten steel affects the wash. In some cases, surface hardness values are obtained on the dried wash before it is used.

Mold washes are being applied to green-sand molds, air-dried molds, skin-dried molds, and dried-sand molds. Practically all oven-dried and skin-dried molds are given a wash. In most cases, the wash is applied before the drying cycle; and in some cases a second coat is applied to the mold after the drying cycle, but while the mold is warm, so that the contained mold heat will thoroughly dry the second mold coat. Green-sand molds are washed only rarely, and these molds are usually for large castings; however, it is known that castings weighing only 35 lb. have required production in washed green-sand molds.

It is the practice in applying wash to molds to be certain that the facing of the mold is moist. The application of a wash to an air-dried face does not result in a good bond between the wash and the mold, with the result that the wash may crack or peel even before the molten metal covers it. Some foundries, in an endeavor to prevent too much air-drying of the mold before the wash is added, resort to the use of a precoat of water or a dilute solution of molasses-water or Glutrin-water just prior to the application of the wash.

Wash is applied on cores before drying or after drying while the core is still hot. In many cases the double coating is used. This procedure requires the core to be washed before baking and then rewashed after baking while the core is still hot. If washed after the core reaches room temperature, the core is returned to the core oven or redried by a torch. Occasionally green-sand cores are washed, but usually these cores are air-dried before using in the mold. In most cases the wash is applied to green-sand molds and cores by spraying.

The atmospheric temperature and degree of humidity are apparently instrumental factors in obtaining good stripping conditions, both from the standpoint of affinity of the mold to these factors and the condition of the metal. Some foundries report changing their facing mixtures and washes at the beginning of the summer and winter seasons. Certain foundries check the temperature and humidity every day and vary their facing mixtures accordingly, especially as to moisture content. The permeability, bond, and moisture content of any facing sand and core sand do not have a very important bearing on the type of wash used. The difference in gravity of the wash and thickness of the coating are of far greater importance.

Tests of mold washes have been made regarding the ease of stripping the mold from the casting.[6] The results seem to indicate that the stripping properties improve with rise in sintering point. Silica-flour-base washes generally produce high sintering points. The base sand on which the wash is applied affects the sintering point after washing. If the sintering point of the base sand is low, it will be raised by a wash but will be still lower than if a higher sintering-point base sand were used.

It is generally agreed that the quantity of clay and organic binders in a wash should be held to as low a percentage as possible, so that a high sintering point can be maintained.

The gas-forming materials should also be maintained at a low figure, since gas expands at the high temperatures—thus requiring molds of a high permeability in order to allow the gas to flow from the mold without setting up high back pressures. Since one of the purposes of a mold wash is to form an impermeable facing to prevent metal penetration, it seems rather incongruous to form an impermeable facing by the use of materials that are in themselves gas forming and whose escape through the impermeable layer must be sought.

In order to obtain the best results from a wash, it is advisable that the wash have approximately the same expansion characteristics under heat conditions as the facing sand. If a wash expanded considerably more than the facing sand, or if the wash were undergoing contraction while the facing sand was expanding, it is easily seen that the wash layer would crack and perhaps spall or flake, thus producing unsightly casting surfaces. A partial remedy for this condition is to apply only a very thin coat of wash. The proper solution of the problem is to make expansion studies of both the wash and the facing sand, and to prepare a wash that will have expansion characteristics similar to those of the facing sand.

One foundry has reported on a comparative study of five mold washes[25] applied in strips on the same casting. The casting was fairly heavy and was poured hot in order to obtain an exaggerated condition. Each strip was gated individually, so as to eliminate temperature differentials in the mold. The following washes were investigated:

1. Silica flour with a resin carrier
2. Silica flour with molasses-water carrier
3. Zirconium silicate
4. Aluminum paint
5. A reducing wash containing graphite

All washes, with the exception of the aluminum paint, increased the sintering point of the surface of the sand. The silica-flour washes gave the greatest increase. Neither the aluminum paint nor the reducing wash showed any improvement over the sand without wash when in contact with the molten metal. The zirconium silicate wash gave slightly the best peel and smoothest surface. The silica-flour washes, however, gave the best all-round results; and the one containing resin showed slightly less spalling and erosion than the one mixed with molasses-water. The casting surface and peel were the same in both silica-flour washes.

A test of 18 mold and core washes was carried on in a steel foundry, on an especially designed test casting weighing 30 lb. All washes were mixed to a consistency of 40° Bé., and both green- and dry-sand molds were

washed.   The castings were examined before and after shot blasting as
to the following: (1) surface condition, (2) erosion at corners, (3) peeling
action of sand, and (4) core-cavity appearance and metal penetration.

The green-sand molds without wash gave about the same general
conditions of acceptability as any of the washes used on the green-sand
molds.   The washed dried-sand molds showed conditions of better
acceptability than the green-sand molds.   This was particularly true of
the core-depression-cavity area.   Metal penetration in the core (casting
depression) was generally greater in the case of green-sand molds.   The
silica-flour wash prepared by the foundry produced results that were
equivalent to any proprietary wash.   However, the silica-flour wash had
little ability to stay in suspension.   Most of the proprietary washes
stayed in suspension very well, and certain of these gave very excellent
results.[40]

Metal Penetration (Burning In).—Most washes used in the steel
foundry give good, but not perfect, results.   Occasionally, the wash seems
to break down.   The mold or core sand does not peel off readily but
adheres to the casting face.   This condition is frequently spoken of as
"burnt-on sand."   The terms "burnt on" and "burning in" imply a heat
effect, fusion, or at least sintering.   In reality, heat, fusion, or sintering
is only one of the causes of adhering sand; the other cause is a mechanical
action that can occur with no heat other than that required to heat the
sand to high temperatures.

Several theories have been proposed as to the reasons for adhering
sand.   Dierker[1] has proposed that the formation of iron silicate is
responsible for adhering sand.   He believes that an iron oxide scale is
formed at the mold face, as a result of oxidation of the solidifying metal
by air and moisture present in the sand.   The ferrous oxide that is formed
is fluid (m.p. 2588°F.) at steel-solidification temperatures, and it reacts
with the silica of the sand grains to form a liquid iron silicate.   Since
ferrous silicate has a melting temperature of about 1200°F., it could
readily penetrate the interstices between the silica grains.

The formation of iron oxide and ferrous silicate and their penetration
into the sand are shown in Fig. 226 by Goodale.[2]   The sketch shows the
mold surface being penetrated by iron oxide and ferrous silicate to an
extent depending upon its porous condition.   The formation of the ferrous
silicate from the casting surface outward into the sand mold makes it
difficult to produce castings with clean surfaces.   The ferrous silicate
penetration theory has not been supported by experimental evidence, and
no one has actually reported the presence of ferrous silicate.   Ferrous
oxide has been reported by analysis, however.

It is reported by Caine[41] that a number of steel foundries have
checked the iron oxide analysis of sands and found that the iron oxide

varied for used sand between 0.5 and 1.0 per cent. The actual content was determined not according to frequency of sand usage, but by the total iron content of the new sand as received. In fact, additions of 0.5 to 5.0 per cent of finely ground iron oxide to the sand showed no lowering of the B sintering point but an increase in the A sintering point. No difference could be seen in the amount of burnt-on sand, or the manner in which the sand peeled from the castings.

The problem of penetration is apparently more complicated than can be explained by the formations of low-temperature-melting products.
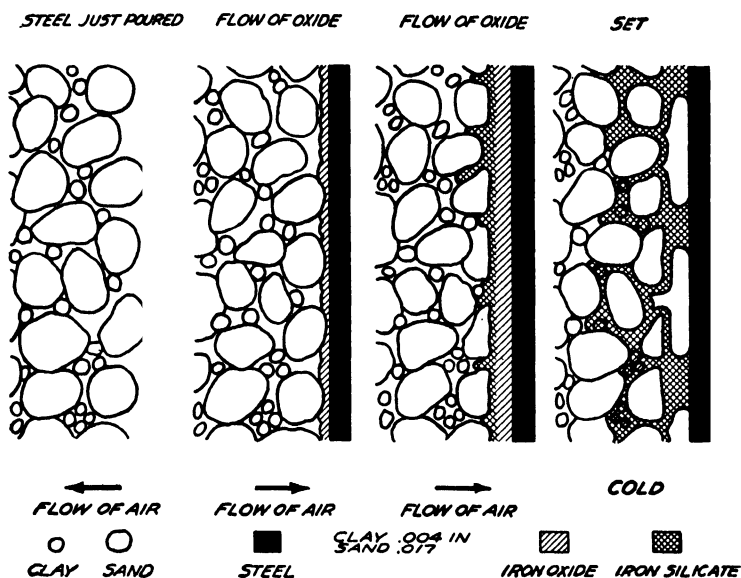
Fig. 226.—Illustrating various stages of penetration of oxide into cores. (*Goodale.*[2])

The burnt-on sand or metal penetration may be considered as of two major causes: fusion, and mechanical penetration.

Any number of experiments have been performed to correlate the behavior of sand under molten-steel action with the fusion of the sand. It was pointed out under Refractoriness that the pyrometric cone test is not applicable to foundry sands. The manner in which fusion is responsible for the burnt-on effect can be best illustrated by a series of microphotographs taken from the work of Caine.[25] Figure 227 is a photomicrograph of the facing sand of a mold as rammed. The sand is a synthetically bonded crude sand of AFA 56 fineness, permeability 120, and B sintering point of 2850°F. The light areas are the individual sand grains, and the dark areas are voids filled with bakelite mounting mediums, holding the sand grains in place for photographing.

Figure 228 shows the sand surface of a 500-lb. casting that peeled

perfectly on shakeout.   The casting was poured at 2900°F.   A comparison of Figs. 227 and 228 will show that nearly all the grains have fused, and only the larger ones remain unaffected by the heat.   Despite this degree of fusion, the sand stripped readily from the casting and cleaned
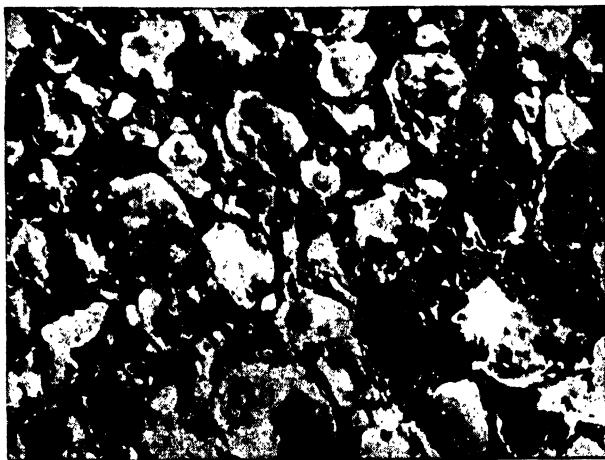
Fig. 227.—Facing sand as rammed.   Oblique illumination.   × 20.   Light areas, sand grains; dark areas, voids.   (*Caine.*[25])
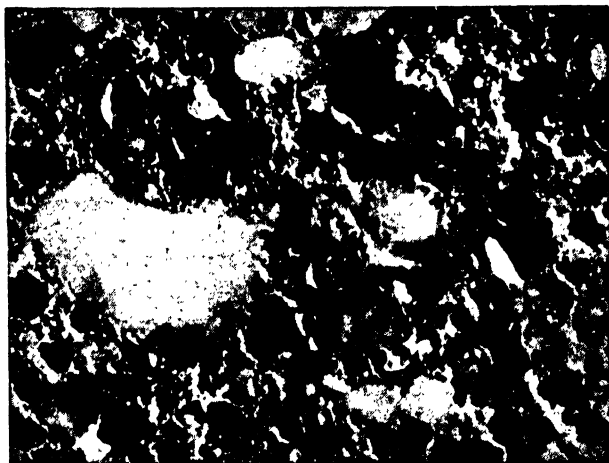
Fig. 228.—Facing sand in contact with molten steel at 2900°F.   × 20.   Dark areas, voids.   (*Caine.*[25])

easily from inside corners by blasting.   Back of the fused area, which was about $\frac{1}{64}$ in. thick, no fusion was evident.

The effect of increasing the pouring temperature to 2940°F. on the sand surface is shown in Fig. 229.   Slightly more fusion is shown than in Fig. 228, and the size of the voids has increased appreciably.   The sand

adhered to the casting on shakeout and was removed with difficulty by blasting. Some metal penetration into the larger voids, as shown by Fig. 230, was responsible for the adhering sand. The increase in pouring temperature of 40°F. was sufficient to open up slightly larger voids in the
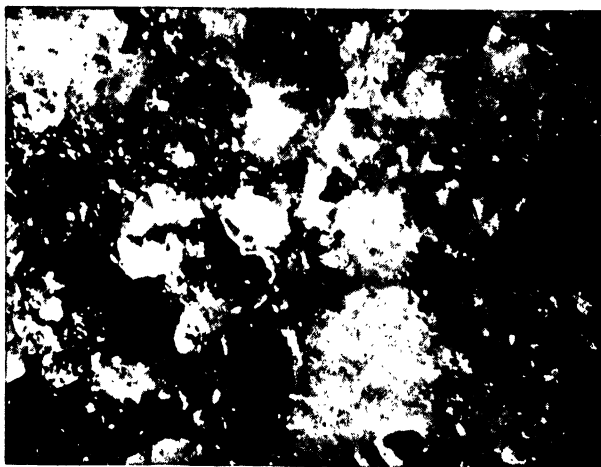
Fig. 229.—Facing sand in contact with molten steel at 2940°F. × 20. (*Caine.*[25])

Fig. 230.—Facing sand in contact with molten steel at 2940°F. Vertical illumination. × 20. Light areas, steel; small voids show dark. (*Caine.*[25])

sand and to allow the steel to penetrate. If the temperature of the metal is further increased, the voids become larger and penetration increases, as can be seen from Fig. 231. In this case complete penetration of the photographic field took place. The sand-metal mixture adhering to the casting is a mechanical mixture of the larger grains of the sand and the

steel. A condition similar to that shown in Fig. 231 is an extremely serious cleaning problem.

The fusion of the smaller grains of sand and the formation of voids, followed by the fusion of larger grains with the formation of a greater number of larger voids, followed by metal penetration, appears to be a satisfactory explanation of the mechanism of the burnt-on condition. According to Caine,[41] the amount of adhering sand can be measured quantitatively by measuring the amount of penetration. When penetration is less than 0.005 in., castings peel perfectly or shake out with a very smooth surface. With penetration from 0.005 to 0.015 in., the castings still peel nicely on shakeout, but the surface is slightly rougher, with a
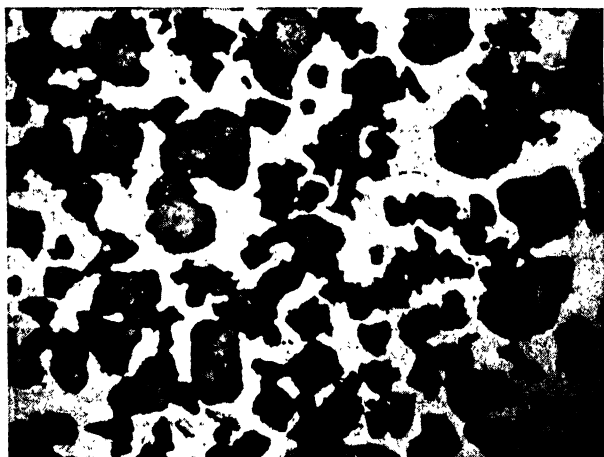
Fig. 231.—Complete penetration of steel between sand grains. Vertical illumination. × 20. Sand grains, dark; steel, light. (*Caine.*[25])

finish that is considered normal for steel castings. As the penetration increases from 0.015 to 0.025 in., the sand adheres to castings on shakeout but is usually removed by blasting and scaling during cleaning and heat treatment. When penetration exceeds 0.025 in., it is usually necessary to chip or grind the sand-metal mixture from the casting. If the penetration exceeds 0.035 in., it is generally cheaper to scrap the casting than to clean it, unless the penetrated areas are localized and the casting is large.

The burnt-on condition can also result from mechanical penetration of metal, as well as from penetration of metal into voids formed by fusion. Mechanical penetration is possible in sands containing large voids caused by too large a grain size or too uniform a grain size in the sand as rammed. Also, veins or cracks in the sand surface permit metal penetration. Soft ramming of molds, low flowability of sands, or any other factor affecting dimensional stability of the mold may bring about

conditions responsible for metal penetration. The presence of large voids can be corrected by selecting a finer sand or by lowering the permeability of the sand through the addition of finer sand to the more open sand. The flowability of the sand may be improved by using a finer sand.

Mold-wall cracking caused by the expansion and contraction of sands permits metal penetration. It has been reported by Woodliff[42] that investigations made on molding sands prone to metal penetration because of mold-wall cracking showed very uniform sands with about 70 per cent on three adjacent sieves. Such sands do not necessarily show a higher expansion than a sand having a wider distribution of grains; but there is indication that the uniformly graded sand will expand more at lower temperatures. This is due, no doubt, to the greater heat conductivity

Fig. 232.—Same as Fig. 230, but perpendicular to the mold-metal interface. Oblique illumination. × 20. (*Caine.*[25])

through the voids in the sand. Surface cracking is observed only a few seconds after specimens of the uniform-graded sands are placed in a furnace temperature of 2500°F. The use of a widely distributed sand will not produce mold cracks due to early expansion.

Occasionally it is found that sand adhered to castings cannot be removed readily by shot blasting, but that the entire sand-steel mixture can be pried off if a lever can be inserted between the sand cake and the casting. This condition results from the steel's penetrating through the mold wash, or a thin layer of facing sand along mold cracks, into a more open sand. The stringers of steel that penetrate the washed surface are not strong enough to withstand any appreciable force; so the entire cake of sand and steel can easily be pried from the casting.

Figure 232, where oblique illumination was used, shows a cross section of the sand-metal interface of a mold washed with a silica-flour-resin wash. The light areas are the fused mold wash and fused sand back of the wash.

Figure 233 shows the same section as Fig. 232 but under vertical illumination.   It can be noted that most of the voids between the sand grains back of the wash are filled with steel.   This particular area does not show a spot where the metal penetrates the wash.   However, such a penetra-

FIG. 233.—Same as Fig. 232, but vertical illumination.   X 20.   White areas are steel.
(*Caine.*[25])

tion of the wash is shown in Fig. 234.   The white areas surrounding the sand grains are steel.   The crack, filled with steel, in the mold wash can be clearly seen.

The grain size of a sand affects the sintering point and steel penetration.   Penetration decreases with decreasing grain size,[41] even though

FIG. 234.—Penetration of steel into facing sand through a crack in the mold wash.   X 6.
(*Courtesy of R. A. Gezelius.*)

the refractoriness as measured by the *B* sintering point is also decreased (Fig. 235).   The finer, less refractory sand has the least penetration and is easily removed from the casting surface.   These studies were made on a high-sintering-point sand.

Another way in which the grain and void size of a molding sand are changed in the foundry is by adding finely divided silica, such as silica flour. The effect of additions of silica flour on the sintering points and
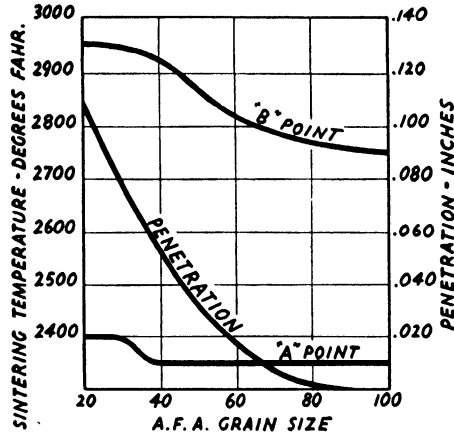
FIG. 235.—Effect of grain size on sintering points and penetration. 4.0 per cent bentonite, 3.0 per cent moisture. Washed silica sand, 98.1 per cent SiO_2. (*Caine.*[41])

penetration is shown in Fig. 236. Caine[41] points out that the *B* sintering point decreases because of the decrease in particle size. The *A* sintering point increases because of the great increase in hot strength owing to the silica-flour additions to the sand. Penetration does not decrease until more than 30 per cent of the mixture is silica flour. In the range of 10
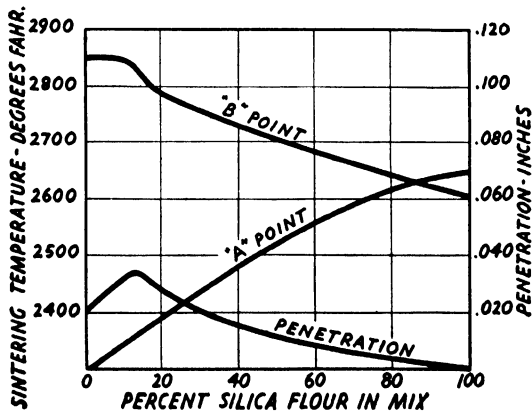
FIG. 236.—Effect of silica flour on sintering points and penetration. Washed silica sand, AFA 59 fineness. 4.0 per cent bentonite, water to temper. (*Caine.*[41])

to 25 per cent silica flour, the small silica-flour particles fuse first and attach themselves to the sand grains, permitting large voids. With large percentages of silica flour the relative number of large grains that form

fusion nucleuses naturally decreases, and, although there is more fusion at the surface in contact with the steel, as shown in Fig. 232, the fusion is uniform and of a glaze type—permitting no penetration but permitting good peel.

The 100 per cent silica-flour conditions are similar to those of a silica-flour-base wash. If the silica-flour wash is applied evenly, it will fuse uniformly and resist penetration of the steel. If the wash cracks or areas of the mold are exposed, the steel may penetrate the wash and the facing sand, either through the conditions of void enlargement caused by fusion or lack of mold stability and high permeability.

Studies by Caine[41] illustrate that by the addition of cereal binders it is possible to decrease penetration and adhering sand. The studies also show that an increase in the pouring temperature, of 60°F., more than doubles the penetration with a corresponding increase in adhering sand.

Small additions of 5 per cent or less of alumina, iron oxide, dolomite, or lime to the AFA clay fraction of a washed silica sand will have but little effect upon penetration and the amount of adhering sand. Additions of slightly over 1 per cent borax are sufficient to cause deep penetration and a severe burnt-on effect. Salt, in amounts up to 5 per cent, actually decreases penetration, even though complete fusion occurs.

Deep metal penetration and considerable adhering sand are found in sands of impure silica. Impurities such as feldspars, when found in the AFA silica fraction in quantities of from 5 to 20 per cent of $CaO + MgO$ or $Na_2O + K_2O$, are responsible for fusion at low temperatures. Steel penetration may be measured in tenths of an inch in such instances.

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CHAPTER XII

CASTING DEFECTS

This chapter describes the appearance, the probable causes, and methods of correcting defects that are found on the surface of steel castings. These defects may be caused by faulty mold or metal conditions or by a combination of both. The defects are listed in alphabetical order and not according to prevalence or degree of seriousness. Defects that are discussed in other sections of this book will be referred to the chapter in which they are more fully considered.

Blows.—A blow usually looks like a shallow depression in the casting surface. Generally the recess has a smooth surface and edge, though in some cases a ragged edge exists. At times the depression has a slight discoloration, this condition developing while the metal is molten. The metal is prevented from reproducing the mold cavity by the presence of mold or core gases, which displace the molten metal. The presence of mold or core gas at the position of the defect results from the trapping of the gas, due to insufficient venting of the mold or cores. The gas originates from organic binders or excess moisture in the sand. Blows may be prevented by increasing the permeability of the sand, by decreasing the moisture content, and by proper venting.

A small, round, and rather deep cavity, expecially at reentrant angles, should not be mistaken for a blow. This defect is probably caused by an atmospheric-pressure break through the casting surface into a hot spot and is associated with shrinkage (see Chap. III).

Buckle.—A depression or break in the casting surface of an irregular shape is known as a "buckle." This defect may be of large, lateral extent and is nearly always found on the cope side. It is similar to a rat tail, or seam, except that it is larger because of a greater breakdown of the mold surface. The defect is the result of radiated heat from the rising metal in the mold, which causes a thin layer of sand at the mold face to expand. The uneven expansion of the thin layer as compared with the sand beneath it results in a cleavage of the sand permitting a dropping of the thin layer of sand. The sand mixture, therefore, has a fairly high, hot expansion. The condition may also be the result of high, hot shrinkage of fines and bonds.

To reduce expansion, an addition of combustible materials should be made, such as cereal binders or the increased addition of clay binders. Expansion may also be reduced by lowering the flowability with the addi-

445

tion of coarse sand grains and increasing the clay content, the green strength, and the green deformation. Sands of excessive strength, resulting from high moisture content, and hard ramming among other things will cause the skin of the mold to dry to a hard, brittle surface, which will easily crack on expansion. The grains of some sands have high expansion and contraction characteristics, and a change to lower expansion and contraction sands is advisable.

Burnt-on Sand.—See Penetration of Metal.

Casting Overweight.—A casting that weighs more than the normal limits allowed is the result of soft ramming of molds or the shifting or misplacement of cores.

Cold Shuts.—Cold shuts or laps appear as lines, irregular in contour and extent, on the casting surface. They seldom appear singly and are more prominent on cope or vertical surfaces. They are caused by metal lacking in fluidity, resulting either from metal low in temperature or the presence of oxide films. They show positions of metal flowing back over solidified surfaces or over colder molten metal covered by thin oxide films. Cold shuts can usually be corrected by pouring with higher temperature metal.

Cuts or Washes.—A cut or a wash appears as an irregular, rough area on the casting surface. In many cases, the sand is embedded in patches or firmly adheres to the metal surface. Cuts are the result of the erosion of the mold surface by the metal flowing over it. The defect can be eliminated by increasing the dry strength of the facing sand. Low hot deformation enhances cuts and washes. In green sand, the defect may be the result of low, or very high, moisture content. Low green strength may also be a contributing cause. Erosion of mold faces may also result from light ramming of the mold or in molds of low mold hardness. Excessive hot spalling of the mold face may also be due to very light ramming of the mold or to low mold hardness. Excessive hot spalling of the mold face is due also to extensive cuts and washes. These may be prevented by the proper ramming of the mold to hardness values of from 80 to 90, and by the increasing of green, dry, and hot strength through moisture and clay additions.

Cracks.—See Hot Tears.

Dirt Spots.—Small spots, ranging from round to irregular shapes, on the casting surface are the result of dirt entering the mold with the metal or eroded from the gate or mold wall by the flowing metal. Slag or refractory material from the ladle or from the gate system may be carried along by the metal and lodged at the mold face. The sand mold may also be eroded in places and the sand carried to other portions of the casting, where it is trapped on a casting surface. Dirt resulting from eroded molds can be eliminated in a manner similar to cuts and washes, by ram-

ming to the proper degree of mold hardness and by increasing the green dry and hot strength of the facing sand. Special care should be given to the maintenance of ladles so that slag and dirt will not enter the molds. Poorly prepared teapot ladles are well-known contributors to dirt in castings.

Drops.—Drops appear as irregular-shaped areas on the casting surface, especially on the cope side. They usually look like enlarged dirt spots and are the result of portions of the mold falling on the rising metal surface. Reflected heat in slowly poured castings, acting on poorly prepared molds, is the principal cause of drops. Low green deformation and low mold hardness are the primary reasons for the collapsing of the mold. These sand properties can be increased by prolonging the mulling time of the sand, making further additions of clay binders or improving the ramming conditions. In some cases, the mold washes break away from the mold face and drop onto the rising metal. This condition results from the application of coatings that are too thick or from lack of penetration of the wash into the facing sand. Drops are commonly spoken of as "burning down of the cope." Practical ways of remedying the condition consist of: (1) faster pouring, (2) a more thorough nailing of the cope surface, (3) an increased use of sand-supporting rods (crabs) in the cope, and (4) strengthening the sand bond by the addition of bentonite or clay.

Fins.—Bits of metal that enter unplanned cavities in the mold are termed "fins." They are usually found at parting lines between flasks or at positions of core prints. Fins result from loosely assembling or fitting together the various components of the mold. Smoothing of mold faces after removal of the pattern, and mismatching of patterns are also responsible for fins.

Hot Tears or Surface Cracks.—Cracks appearing on the surface of castings may be of two types: (1) cold cracks, or (2) hot tears. Cold cracks are straight-line cracks of few branches. They result from large inherent contraction stresses, produced during cooling from the solidification temperature to atmospheric temperatures. These cracks form at comparatively low temperatures (300 to 0°F.). For a complete discussion of cold cracks, see Chap. VIII. Hot-tear cracks have an irregular and jagged appearance, with an oxidized fracture surface. The crack first happens on the surface of the casting and proceeds inward. The crack is wider at the surface, is uninterrupted in depth, and in some cases may traverse the entire section. Stresses active in the formation of hot tears appear to be tensile or shear stresses. These stresses arise from hindered-contraction causes, resulting from mold conditions or casting design. See Chap. IX for a complete discussion of hot tears.

Misruns.—Misrun castings are failures of the metal to reproduce mold contours; they result when the molten metal does not completely fill the

mold cavity. This is primarily caused by low fluidity, and the simplest way of increasing the fluidity of steel is to increase the pouring temperature. However, steel with good fluidity does not readily fill sections of less than $\frac{3}{16}$ in. in thickness, especially if the extent of the section is more than a few inches.

Penetration.—A rough casting surface results when molten metal penetrates into the sand mold. This penetration may extend to a depth of only two or three sand grains, or it may be more pronounced. Penetration of steel into the mold may result from the high permeability of the mold owing to the use of coarse sand, or from low mold hardness caused by soft ramming of the mold. The defect may be corrected by using mold washes or a finer sand of lower permeability. For a more extensive discussion, see Chap. XI.

Pinholes.—Pinholes are small (pin-point size), elongated, smooth-walled gas holes, which occur immediately under the skin of a casting and are brought to the surface after heat-treatment, owing to the loss of oxide scale. Gases in the metal and high moisture conditions in the mold are primarily responsible for pinholes. See Chap. III for a complete discussion of pinholes.

Rat Tail.—A long, irregular depression of little width in the casting surface is called a "rat tail." It sometimes has the appearance of a fissure or a crack, but it is not formed because of stress action on the casting. The rat-tail defect is caused by high-expansion spalling of the mold wall. High mold hardness and an inferior type of facing sand may also cause rat tails. The defect may be remedied by reducing expansion and spalling of the facing sand and by increasing the clay content to produce a sand of greater strength and deformation. An addition of cereal binders to the mix will reduce the expansion of the sand. A reduction of the degree of ramming will lower the mold hardness.

Rough Casting Surfaces.—See Penetration.

Sags.—Changes in the appearance of the desired casting, either as to wall section or general design, are known as "sags." Shifting chaplets or poorly positioned core prints may be responsible for the sagging of cores within the mold, which would alter the appearance of the casting. The mold itself, if not properly supported by the use of arbors, molding rods, or crabs, may settle and produce a sag.

Scab.—A scab is an irregularly shaped mass of metal that usually lies above the casting surface. The defect is due to a combination of spalling and high hot shrinkage of the sand. High hot shrinkage can be retarded by increasing the refractoriness of the sand or by increasing the grain size and reducing the fines in a sand. An addition of cereal binder to the sand will help to reduce excessive spalling. It has also been stated that high moisture content of sands and moisture in cores are primary causes of

scabs. Scabbing can be eliminated in a practical manner by faster pouring of the metal, increased nailing of the cope, and prevention of reflected heat on the cope by shading it with internal chills.

Seams.—A seam is similar to a rat tail, except that it is broader and more confined in appearance. The defect is caused by spalling the cope surface of a mold (see Rat Tail).

Slag Spots.—See Dirt Spots.

Swells.—Metal lumps or bulges on the casting surface or on the walls of inner passage faces are called "swells." They are formed when the mold settles over a small localized area, or when the metal, because of its pressure or weight, compresses the mold in a localized spot. The reason for their formation is that the sand has not been properly rammed in the particular area of the swell, or that, because of mold-relieving materials or conditions close to the mold-metal interface, the weight of the metal has been able to compress the mold at this point.

Thick-thin Sections.—The term "thick-thin sections" refers to a casting defect where one section of the casting is thinner than that of the pattern while a corresponding section is thicker than designed. This condition is caused by the misplacement of cores or the shifting of cores either before or during the pouring of the casting. It can be corrected by careful inspection and dimensional checking of molds just prior to their closing. Loosely set cores are discovered at this time.

Veining.—Veining appears on the casting surface in the form of small, raised lines of metal, similar to veins standing out on parts of the body. They are formed when metal enters the cracks in the facing sand. These cracks are the result of rapid collapsibility of the sand, or high expansion characteristics. This condition may be retarded by the addition of silica flour or iron oxide to the facing sand, and also by increasing the green and dry strength of the sand. The addition of fines decreases the permeability of the sand, which is also helpful in preventing the defect.

CHAPTER XIII

CLEANING, CHIPPING, AND GRINDING OF STEEL CASTINGS

CLEANING

One of the major problems in the manufacture of steel castings is to clean them and to remove excess metal. It is not the purpose of this chapter to discuss the mechanical features of various pressure-blasting, chipping, and grinding units or to consider in detail the various types of equipment available to do the work required in the cleaning department. The plan is to present pertinent information regarding the technical features in the operation control of blast cleaning and grinding.

For many years the rattling and tumbling of castings were accepted methods of cleaning, although by the present standards of surface acceptability these methods of cleaning were not sufficient. Tumbling is used today in the modern steel foundry, but the modern blast-cleaning units are designed to use tumbling in conjunction with compressed air or mechanical blast on the work.

Today, abrasive-impact cleaning may be classified into two main groups: (1) pressure-blast cleaning and (2) mechanical-impact cleaning (see Fig. 237). Pressure-blast cleaning comprises (1) compressed-air-blast cleaning and (2) water blasting. Compressed-air-blast cleaning can be divided into two parts: (1) the direct-pressure principle, and (2) the injector principle. The injector principle includes both suction and gravity feeds.

Suction-feed Method.—In the suction-feed method of compressed-air-blast cleaning, the abrasive is drawn by means of a compressed-air jet from a feedbox into the injector chamber directly behind the nozzle. The accelerated flow of the abrasive is so rapid that once it passes the tip of the nozzle, the air expansion, being unrestricted, drives the abrasive against the work surface at a velocity of about 7,000 to 8,000 ft. per min. at about 90 lb. air pressure.

The main advantage of the suction-feed principle is its simplicity, as is evident from the drawing of Fig. 238. The abrasive is conveyed, for a considerable distance if necessary, by an air current induced in a closed, injector-type gun. This permits locating the abrasive storage bin at any convenient point above or below the point of final delivery. Generally, the equipment is on a lower level than the gun.

A disadvantage is that a smaller volume of abrasive is fed by this type

than by the others, for so much of the air effort is expended in elevating the abrasive to the nozzle. Also, the system generally lacks means for cleaning the continuously used abrasive, which becomes contaminated
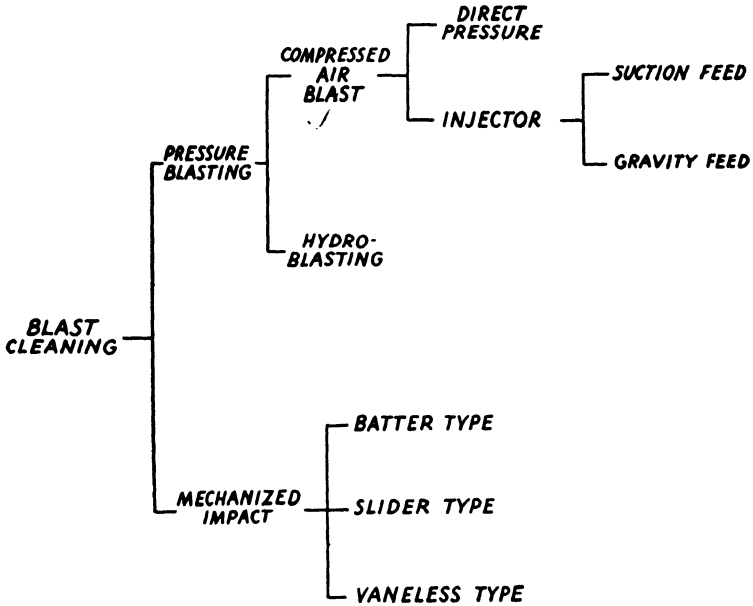
FIG. 237.—Blast-cleaning methods.
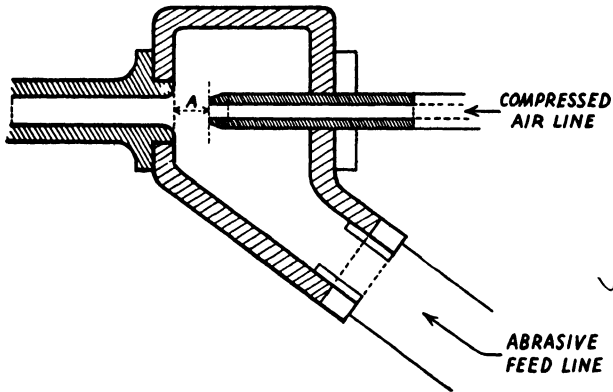
FIG. 238.—Suction-feed blasting.

with sand and scale. Such contamination of the abrasive results in a loss of its cutting efficiency.

In a properly designed gun a static vacuum (suction) of 17- to 19-in. mercury gauge can be obtained. The maximum suction is obtained by adjusting the distance A in Fig. 238. An 80-lb. air pressure should produce a 14-lb. suction.

Gravity-feed Method.—In the gravity-feed method the abrasive is elevated by mechanical means and is fed downward by gravity into the injection chamber behind the nozzle. An orifice plate is usually provided in the feed line to govern the rate of abrasive feed. The average abrasive velocity in this type ranges between 8,000 and 10,000 ft. per min. at about 75 to 90 lb. air pressure.

The simplicity of the gravity-feed arrangement recommends it in many cases, particularly for multinozzle machines such as rotary tables and special applications where final cost, simplicity, and low maintenance are of prime importance. Generally in gravity-feed blasting the gun
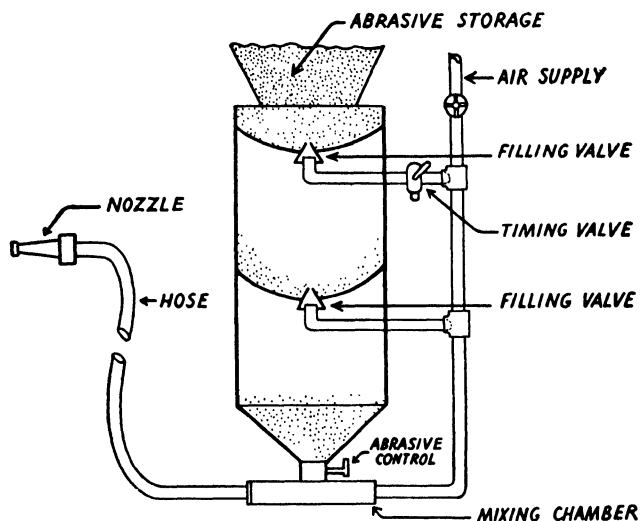
Fig. 239.—Direct-pressure blasting.

points straight downward or at a steep angle. Both gravity-feed and suction-feed blasting are used in applications where the intense cleaning action obtained by direct-pressure blasting is not necessary.

Direct-pressure Blasting.—The direct-pressure principle employs a full pressure of compressed air from the mixing tank to the nozzle. The mixture of compressed air and abrasive is carried to the blast nozzle in a high-pressure pipe line from a pressure tank with integral mixing chambers. This method of air-pressure blasting produces a continuous blasting operation with the maximum of abrasive velocity. The pressure tank is usually separated into two chambers, each fitted with an individual valve (Fig. 239). In the actual operation, the upper chamber of the tank is filled while the lower chamber is discharging into the nozzle line in a continuous stream. The velocity of the abrasive from the nozzle is about 15,000 ft. per min. average.

In this system it is important to regulate the abrasive flow according

to the size of nozzle and the type of abrasive used. These values have not been determined exactly, but it has been found that the orifice between the tank and the abrasive line should be larger than the nozzle area.

Most of the compressed-air-impact cleaning is carried on in a large room such as shown in Fig. 240, where both the operator and the work are

FIG. 240.—Room air blast.

in the same enclosure. In some cases, operators are outside the room and are able to operate the nozzles in the blast-cleaning chamber.

Process of Compressed-air-impact Cleaning.—The impact-cleaning operation is essentially a bombarding of the casting surface by a continuous impact of flying abrasives. An estimate of the number of grains propelled against a surface has been made by Rosenberger.[1] For a $\frac{3}{8}$-in.-diameter nozzle that discharges 12 cu. ft. per min., a total of about 12 million grains of abrasive will bombard the casting per min.

Now, in order for the abrasive to remove the casting surface, or to remove fixed material on the casting surface, the stress exerted by the

flying abrasive particle at the point of impact must exceed the strength of the material at the casting surface.   Four factors determine this:[1]

1. The energy contained in the flying grain
2. The area upon which this energy is expended
3. The strength of the object struck
4. The strength of the abrasive particle

Naturally, the energy contained in one single grain of abrasive is small, but it is because this energy is expended upon an exceedingly small area that a visible destruction of the surface is possible.   Rosenberger[1] takes into consideration the mass of the individual abrasive grain and the area and depth of contact and computes the pressure exerted per square inch as over 10 million lb.   Since the compression
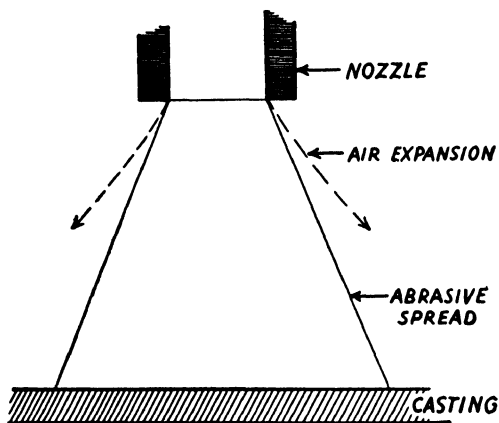
FIG. 241.—Air and abrasive spread in compressed-air blasting.   (*Rosenberger.*[1])

strength of alloyed cast steel does not usually exceed 150,000 p.s.i., it will be seen why blast cleaning is so tremendously destructive.

There are several items that should be kept in mind regarding the velocity and flow of abrasives through nozzles.   In the first place, as the orifice of a nozzle increases in size because of wear, an increase of the air flow results.   An increase of only $\frac{1}{16}$ in. in the diameter of the nozzle orifice will increase the air flow about 35 per cent.   This is a point that should receive careful consideration, particularly when selecting air compressors for blasting equipment using a multiplicity of nozzles.

It is possible to offset the loss in abrasive striking power when increasing air flow demands cannot be met by increasing the flow of abrasive, since in the pressure line the abrasive has a much slower velocity than the air and therefore has a retarding effect upon the air flow.   In general, pressures should range from 80 to 100 lb. for compressed-air-blast cleaning of steel castings.

The higher the average abrasive velocity through a nozzle, the more

rapid the wear. Nozzles wear least when shot is used and wear fastest with artificial abrasives such as silicon carbide particles. Also, the abrasive velocity is related to the specific gravity of the abrasive material, being lowest when the specific gravity is the highest. The shape of the abrasive particles also influences the velocity in that the spherical particles travel slower than angular particles.

Air, in issuing from a nozzle, expands into the atmosphere in some fashion such as may be indicated in Fig. 241. This expansion exerts a spreading effect upon the abrasive; and, of course, the lighter and finer the abrasive, the greater will be the spread. Theoretically the inverse-
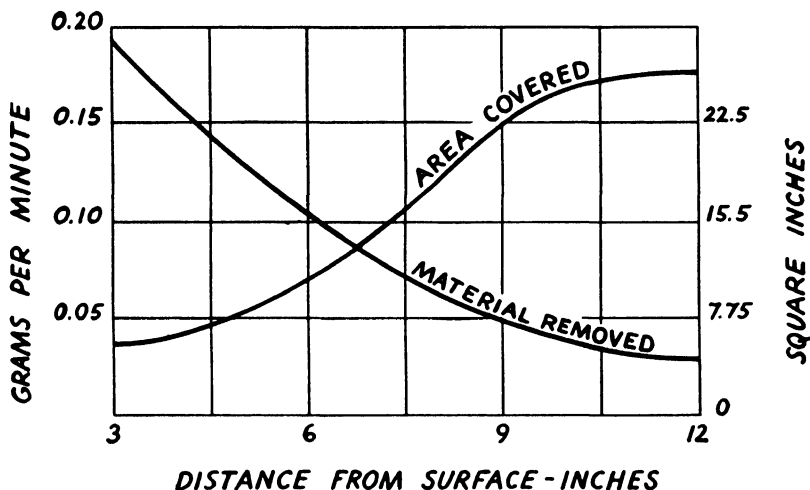
Fig. 242.—Effect of varying distance on surface area blasted. (*Bradshaw.*[3])

square law applies to the surface effectively covered, since the closer the casting is to the nozzle the smaller will be the diameter D of the blasted area. If the nozzle is removed a distance from the casting so that the diameter of the blasted area is $2D$, the area covered is four times as great.

The most effective blast cleaning is done when the casting is perpendicular to the nozzle. If the casting is presented at an angle to the impinging abrasive stream, a less intense abrading action results. Figure 242 shows the effect of varying the distance from the surface treated on the abrading action. Reams[2] recommends that the abrasive size should not exceed one-third the size of the nozzle orifice for direct-pressure blasting.

Mechanical-impact Cleaning.—In the mechanical-impact method of cleaning the abrasive is projected against the casting at a high peripheral speed by centrifugal force. The resultant velocity of the abrasive is equal to a component of the radial and tangential forces. The centrifugal-

wheel unit is a high-speed revolvable radial wheel operating at 1,800 to 2,500 r.p.m.

Abrasive propelling wheels have been divided into three general classes:[1] batter-type wheels, slider-type wheels, and vaneless wheels. In the batter-type wheels the abrasive is fed to the batting vanes either radially or axially in such a manner that when it contacts the vanes, it has practically no velocity in the direction of the vane travel and is, there-
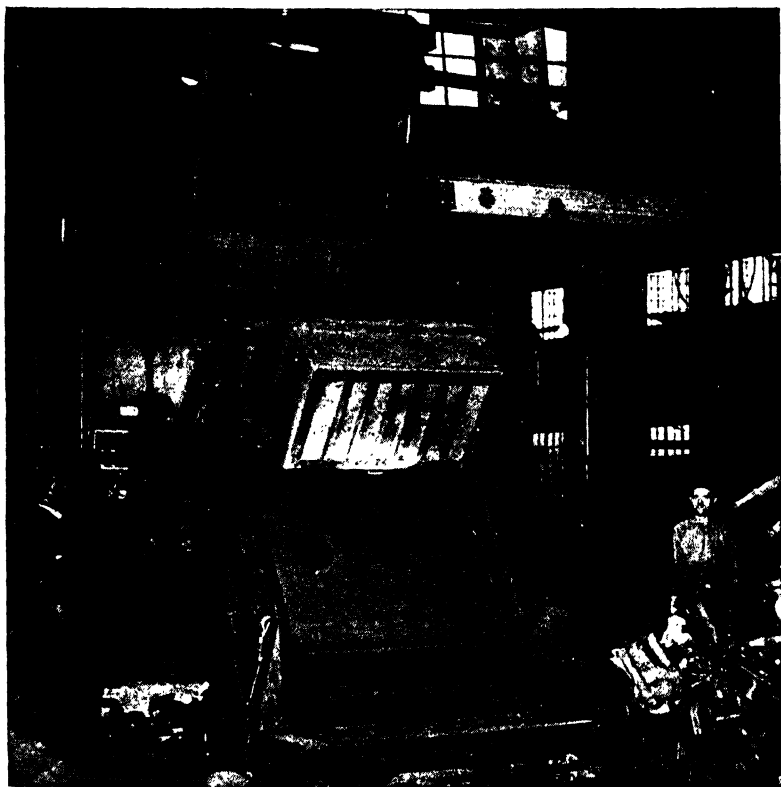
Fig. 243.—Mechanical impact, cleaning-barrel type.

fore, hit or batted by the vane with considerable impact; in other words, the abrasive particles obtain their velocity by impact.

The slider-type wheel provides for the abrasive to be deposited, by some mechanical means, on the innermost edge of the propelling vane. The abrasive slides outward along the vane until it reaches the tip where it is thrown off by centrifugal force. The abrasive particles are given both radial and circumferential velocity.

The vaneless type provides for the abrasive to be spilled into the inside rim of a high-speed revolving concave wheel where it is held in place by

centrifugal action and a disk ring. The abrasive particles slide through ports in the ring on to a wear-resistant surface, where they are swept off by a freely rotating steel disk.

The batter-type wheels are widely used in Europe, but they have met with little favor in the United States. This is due largely to production requirements of handling large quantities of castings in a short period of time.

The vaneless type is the latest development and has been developed because of the rapid wearing of the propelling vanes in the other types. Since the abrasive so thrown has no radial velocity, this type of wheel

FIG. 244.—Mechanical-impact multispindle cleaning table.

should run 30 per cent faster than the standard-type vaned wheel of the same diameter. The wheel pattern is long and narrow, 2 to 2½ in. wide, and uniform.

Most commercial installations of mechanical-impact machines are of the slider type. By allowing the abrasive to be deposited on the inner end of the vanes at a definite point in relation to the circumference of the revolving wheel, it is possible to obtain a directional discharge of the abrasive. The centrifugally projected abrasive stream cannot be aimed at a specific target such as can be done with a compressed-air-nozzle blast stream. For this reason the mechanical-impact cleaning has found its best application in connection with the use of tumbling barrels, where all faces of the castings are presented to the abrasive stream (Fig. 243). Multispindle cleaning tables are also used (Fig. 244). '

The centrifugally projected blast stream is approximately 30 in. long and 6½ in. wide, with a zone of blast intensity of 12 by 4 in.[2]   By revolving or tumbling the castings, it is possible to compensate for the lack of uniformity of the blast stream so that the castings are uniformly cleaned.

In mechanical-impact blasting the abrasive weight or mass does not have the effect on its velocity as in compressed-air blasting.   The velocity of the abrasive, as it is discharged toward the casting to be cleaned, is at the rate of about 200 to 240 ft. per sec.   The impact of the abrasive projected centrifugally against the casting surface varies with the wheel speeds, but, according to Reams,[2] it averages about 15 to 18 lb.   This is considerably less than the estimated 40 lb.[1] for the normally used compressed-air blasting of steel castings.   This seemingly light impact against the casting is compensated for by the large quantity of abrasive particles projected against the castings.   The peening action of the stream is not so severe as with compressed air, and cracks or flaws in the casting surface are left open, making it easier to inspect and repair casting surfaces.   It has also been noted that the finer the abrasive, the faster is the cleaning action.

The present trend in the steel casting industry is that there are more installations of mechanical-impact equipment than of air-blast cleaning. This speaks well for the economy of centrifugal-blast equipment.   The reasons for this trend are

1. Air compressors, their purchase, operation, and maintenance can be eliminated.

2. Mechanical-impact equipment is self-contained and self-sufficient.

3. Floor space is saved.

4. Incidental losses such as air, leaks in pipe lines, valves, etc., are eliminated.

5. Moisture trouble in compressed-air equipment is absent.

6. Mechanical-impact equipment is always ready to run.

The disadvantages of mechanical-impact equipment are as follows:

1. The lack of flexibility of wheel application and its inability to produce a highly concentrated stream of abrasive comparable to that issuing from a nozzle.

2. Lack of uniformity of the intensity of the abrasive stream.

3. Short life of vanes, impellers, and cases, the changing of which is not done easily or so quickly as changing air-blast nozzles.

Hydroblast.—The cleaning of castings, particularly the removing of sand and cores, by water pressure has been carried on with some success for the last 20 to 25 years.   Many of the attempts to develop a wet-cleaning system met with indifferent success owing to the fact that they were low-pressure systems using no abrasive in the water stream.   Water alone at low pressure does not effectively clean or wet the dust.

Water has a carrying power in weight and size of material that increases with sixth power of the velocity. Therefore, a stream of water at very high velocities will carry large amounts of abrasives and will impart velocity to this material. Water alone without abrasives will remove sand from castings, but when solids are added to high-velocity streams of water, the erosive action is increased materially. The relative efficiency. of a stream of water and sand as compared with water alone at the same pressure has been determined by Webster[4] to be about 7 to 1.

In the present system of hydroblast cleaning a stream of approximately three parts water and one part sand is projected through a gun against the casting at a pressure of 1,200 lb. with a velocity in excess of 18,000 ft. per min. About 28 gal. of water are used per min. through a 0.190-in.-diameter orifice, and the water carries from 60 to 70 lb. of sand per min.

The pressure gun is of the injector-suction feed type as shown in Fig. 238. A vacuum of between 27 and 28 in. of mercury is maintained at the mixing-chamber end of the sand line. The suction-feed principle is used in preference to the gravity feed, since it has been found that there is more tendency for the feed line to clear itself if plugging of the line by the sand takes place. It is essential that wet sand be used in order to obtain proper functioning of the gun. A definite head of water above the sand in the sand tank is usually maintained to assist in the feeding operation.

Studies by Webster[4] showed that there seemed to be little difference between the use of coarse and fine sands as the abrasive, and that no special abrasive sand is required in the operation. In fact, the sand that is cleaned from the casting is usually entirely suitable for casting cleaning.

The effectiveness of sand and water does not change significantly up to distances of 36 in. from nozzle to the casting. However, at distances greater than this the stream becomes less concentrated. It has been suggested[4] that the most effective angle for the water and sand stream to impinge on the casting is 45 deg. The stream is most effective when there is no interference through rebound of the sand particles striking the surface of the casting being cleaned. Steel shot is not often used as an abrasive in hydroblasting since the shot would be mixed with sand that is usually recovered.

An advantage of hydroblast cleaning is that the high velocity of the water stream wets out any dust at the source, and it removes dust created by other foundry operations and left suspended in the air. The castings are cleaned in a special room, which usually requires no ceiling and permits full crane service on heavy work, or in tumbler barrels for smaller work (see Fig. 245). The removed sand and waste water pass through the floor of the room, flow over a traveling vibrating screen for removal of

all cinders, nails, and oversize material, and collect in a sump tank. The sand and water are pumped from the tank to counterflow classifiers. Burned sand, clay, organic matter, and fines are removed from the sand by the classifiers and pass into a sludge tank for dewatering and disposal. The usable sand from the classifiers passes into the blast-sand tanks to be used again in hydroblast cleaning. Sand in excess of that needed for hydroblast cleaning is diverted to dewatering boxes for reuse as core or facing sand, after drying to the proper moisture content.

FIG. 245.—Hydroblast cleaning of a steel casting.

The hydroblast method of cleaning is a very effective way of removing cores, for the velocity of the water assists in washing out the baked sand.

Abrasives.—The abrasives mostly used for cleaning steel castings are grit, shot, and sand. The grit and shot are commonly known as "steel abrasives," although they are really chilled cast iron of high hardness of approximately 800 Brinell. The shot is prepared by allowing a steam jet to impinge on small streams of molten alloy cast iron, thereby breaking it up into globules, which are suddenly solidified in tanks of cold water. The shot is then passed over grading screens. Steel grit is made from steel shot by crushing and grading it.

Shot is spherical shaped and can be purchased in sizes varying from less than $\frac{1}{64}$ in. in diameter to sizes of $\frac{1}{8}$ in. or over. Grit can be purchased in sizes resembling dust to granules measuring up to $\frac{1}{8}$ in. or over

in one or more directions. They are graded by the manufacturers by an arbitrary number having no direct relation either to size or weight. Each manufacturer has his own numbering system. Comparative grit and shot numbers are given in Tables LXVII and LXVIII. Since the shot particle is spherical, it will weigh more than twice as much as the grit particle of the same sieve size, and it is therefore particularly adapted for work requiring peening. The removal of scale resulting from heat-treatment and the removal of large cores are more effective with shot blasting than with grit cleaning.

TABLE LXVII.—COMPARATIVE SIZES OF METALLIC GRIT BY DIFFERENT MANUFACTURERS

Pangborn Corp., Hagerstown, Md.	American Steel Abrasive Co., Galion, Ohio	Globe Steel Abrasive Co., Mansfield, Ohio	Pittsburgh Crushed Steel Co., Pittsburgh, Pa.	Steel Shot & Grit Co., Boston, Mass.	Steelblast Abrasive Co., Cleveland, Ohio	National Steel Abrasive Co., Cleveland, Ohio	Alloy Metal Abrasive Co., Adrian, Mich.
No.	No.	No.	No.	No.	No.	No.	No.
12	8	8	12	10	8	..	10
14	10	12	14	11	10	10	12
20	12	14	20	13	14	12	14
30 Reg.	14 & 18	20 & 30	30 Reg.	16 Reg.			
30C	14	20	30C	16C	20	14	20
30F	18	30	30F	16F	25	18	25
40	38	40	40	30	28	38	30
50	45	50	50	40	35	..	40
60	60	60	60	60	45	45	50
70	70	70	70	70	50	60	60
90	90	90	90	90	80	70	70
120	120	120	120	100	120	80	90
170	170	170	170	....	...	90	

The grit abrasive, being more angular, acts like countless minute chisels. The steel grit is better adapted for casting surfaces where surface imperfections should be exposed for ready inspection and repair and where a mat finish is wanted. One ton of steel abrasive will do approximately the same amount of work as 40 tons of sand.[1] The advantage that comes in handling, storing, trucking, and freight expenses is a considerable reduction for handling the smaller quantity of abrasive.

The best quality of chilled-cast-iron abrasive should be used, since it lasts longer and the life of the wearable parts of the blast machine is shortened greatly by the use of inferior-grade abrasives. In specifying

uniform size of shot or grit, the following tolerances have been advocated by Reams:[2]

Five per cent is retained on the screen of the desired opening.

Ninety per cent passes the screen of the desired opening and is retained on the screen of the next smaller size opening.

Five per cent passes through the retaining screen. With such a classification, Table LXIX lists recommendations for shot and grit sizes for steel castings.

TABLE LXVIII.—COMPARATIVE SIZES OF METALLIC SHOT BY DIFFERENT MANUFACTURERS

Pang-born Corp., Hagers-town, Md.	American Steel Abrasive Co., Galion, Ohio	Globe Steel Abrasive Co., Mans-field, Ohio	Pitts-burgh Crushed Steel Co., Pitts-burgh, Pa.	Steel Shot & Grit Co., Boston, Mass.	Steel-blast Abrasive Co., Cleve-land, Ohio	National Steel Abrasive Co., Cleve-land, Ohio	Alloy Metal Abrasive Co., Adrian, Mich.
No.	No.	No.	No.	No.	No.	No.	No.
5	12C	1C	5	5	..	...	5
6	13C	1	6	6	6	11	6
7	13	2	7	7	7	12	6½
9C	14C	2½	9C	9C	8	13C	7
9F	14F	....	9F	9F	..	13	8
10C	15 by 16	3C	10	10C	9	14	9
10F	16	3	....	10F	10	15	10
12	17	3½	12	12	..	17	12
14	19	12	14	14	14	...	14
16	20	4	16	16	20	19	16
20	22	5	20	20	22	20	20
26	25	5½	26	26	30	22	22
30	28	6	30	30	35	25	25
40	30	7	40	40	40	28	30
					50	30	

The mechanical-impact machine can operate with a greater flow of abrasive than can the pressure-blast machines. In order to increase the flow of grit in the mechanical-impact machines, an addition of a small quantity of shot is necessary.

Peak metal-cleaning efficiency from abrasive-blast-cleaning equipment (mechanical or pressure) depends upon keeping the metallic abrasive clean. Molding sand, scale, and broken abrasive fines retard cleaning action and increase the rate of wear in the blasting machine. Failure to keep foreign material out of the abrasive can lower cleaning speed as much as 50 per cent. A ventilating and dust-collecting system of ade-

quate capacity and correct design will maintain cleaning efficiency at a high level, if the system is regularly inspected to ensure its operating as recommended. Foundries should from time to time make operating-efficiency tests on abrasives used. In this way the correct abrasive size can be selected for the type of castings being cleaned, and an estimate of the quality of the abrasive can be obtained. To determine the relative

TABLE LXIX.—ABRASIVE SELECTION FOR GENERAL BLAST CLEANING OF CASTINGS Reams[2]

Steel grit

Steel casting	Screen opening, in.					
	0.075	0.065	0.053	0.044	0.032	0.0022
Large........................	Use	Use				
Medium......................	....	Use	Use	Use	Use	
Small, commercial finish........	....	Use	Use	Use	Use	
Small, fine finish..............	....	....	Use	Use	Use	Use

Steel shot

Steel casting	Screen opening, in.			
	0.055	0.046	0.031	0.024
*Large...	....	Use	Use	
*Medium..	....	Use	Use	
Small, commercial finish.........................	....	Use	Use	
Small, fine finish with deep core pockets..........	Use	Use	Use	Use

* Use mostly to remove scale due to heat-treating.

efficiencies of the abrasives tested, the following operations should be checked:

1. Abrasive breakdown

Samples of the abrasive should be analyzed by screening at 4-hr. intervals. The condition of the abrasives should be noted as to the quantity of usable material remaining in each sample.

2. Cleaning efficiency

The time required to clean identical loads of work should be ascertained. The castings must be comparatively equally cleaned, in order that there may be a fair basis of comparison.

The circulating stream of metallic abrasive as used in a blast-cleaning system undergoes disintegration into particles. The nature of this breakdown is shown by Hurst and Todd[5] in Table LXX.

From Table LXX it will be noted that the particles below 40-mesh are extracted almost completely by the dust-extraction system. The grit remaining in circulation showed considerable wear of the sharp edges, but in spite of this the grain always remained rough and reasonably angular.

TABLE LXX.—DISINTEGRATION OF No. 16 METALLIC GRIT IN ROOM-TYPE
PRESSURE BLASTING
HURST AND TODD[5]

Elapsed opera-tion time, min.	Percentage remaining in sieve						
	No. 16	No. 18	No. 24	No. 30	No. 40	No. 50	No. 60 and −60
Original	40	14	40	4	2		
10	35	9	35	9	7	3	2
20	31	10	40	11	6	1	1
30	25	10	45	11	5	3	1
40	15	7	40	20	13	3	2

Sand as an abrasive for steel-casting cleaning is being used less and less as the years advance, and its greatest application is for use in hydroblast equipment. When sand is projected under high velocity against a hard surface the sand grain splits along the cleavage planes of the grains and only a few (four to six) such impacts are necessary to reduce any sand grain to fine particles that have a negligible blasting effect. Steel abrasives, on the other hand, are solid, homogeneous masses devoid of cleavage planes and can be used over and over again since they disintegrate slowly. Typical screen analyses of two sandblast sands are shown in Table LXXI.

TABLE LXXI.—TYPICAL SANDS FOR PRESSURE-BLAST CLEANING
Great Lakes Foundry Sand Company

Screen, U.S.	20	30	40	50	70	100	−100
Diamond	0.9	39.7	55.3	3.0	0.5	0.5	0
Gold jacket	9.1	43.3	42.3	4.3	0.5	0.1	0.2

Other abrasives, such as silicon carbide and manufactured alumina, have been found satisfactory for special blast-cleaning applications. Silicon carbide is very hard, although brittle. Manufactured alumina is hard but comparatively expensive. These materials are seldom used for cleaning steel castings.

Nozzle Wear.—One or two points should be made regarding nozzle wear. Experience has proved that the useful life of a nozzle is proportion-

ate to the specific gravity and the fineness of the abrasive used, and to the blast pressure. Spherically shaped abrasives will cause less wear than those that are angularly shaped, and the finer the abrasive used, the faster becomes the nozzle wear. Wear also increases with the increase in blast pressure.

A few uncorrelated studies have been made in the steel foundry on nozzle wear. One of two foundries using the same make of nozzles lined with tungsten carbide reported a life of 2,500 hr. with steel shot, while the other foundry reported a life of only 400 hr. with steel shot. The
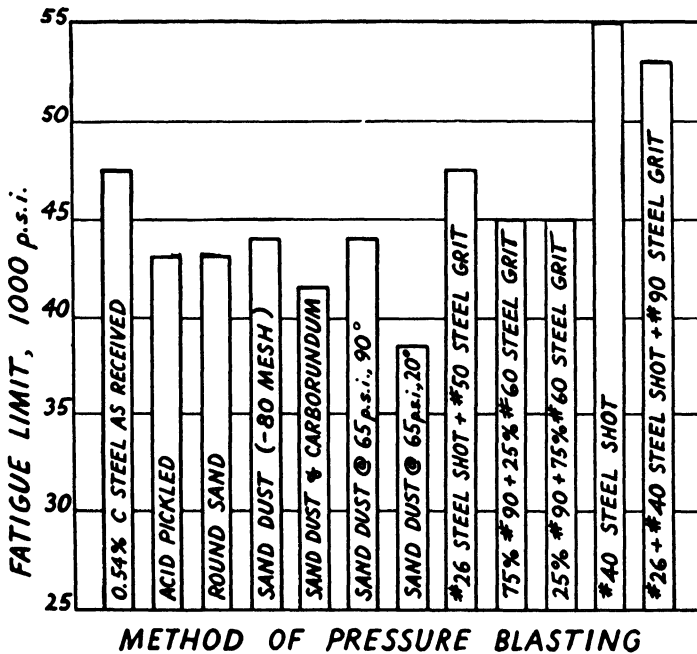
FIG. 246.—The fatigue limit of SAE 1050 steel as affected by some cleaning methods. (*Frye and Kehl.*[7])

wide variation of answers suggests that a more careful study be made of the variables.

Shot-blasting Effect on Fatigue Life.—There has been some discussion from time to time of the fatigue resistance of steel as influenced by the various cleaning methods. Some information along these lines has been published by Frye and Kehl,[7] and also by Zimmerli.[8] Their findings may, or again they may not, apply to the cleaning of steel castings. It is possible that if cleaning by blasting or pickling is the final operation, the results may apply.

It appears that blasting produces two surface effects, *viz.*, notches and cold work. The first is deleterious; the second is beneficial within

limits.   Blasting steel grit, shot, or mixtures of the two at pressures of
about 80 lb. is definitely superior to sandblasting or the usual sulphuric-
acid pickle in its effect on the fatigue limit.   In fact, such blasting
improves the fatigue limit, as is shown in Fig. 246.

The angle of blastings is important.   The best results at the high
pressures for maximum fatigue life are obtained when the angle of blast-
ing is about 75 deg.   It is also evident that the larger the size of steel
shot used, the greater will be the amount of cold work.

Acid pickling definitely lowers the fatigue resistance of steel since it
produces surface pitting of the steel, which reduces the fatigue strength.
It is pointed out by Zimmerli[8] that the increase in fatigue life of steel
by shot blasting is a surface phenomenon that can be removed by heating
at temperatures as low as 800°F.

Tumbling.—Rattling or tumbling is the oldest form of machine clean-
ing of castings.   This type of cleaning is being used in some steel foundries
today, though no new equipment of this type is being purchased.   The
cleaning room during the Second World War was such a bottleneck of
production that every piece of cleaning equipment was put into service,
even to the reinstalling of obsolete tumbling equipment.

Castings are tumbled by placing them in a rotating barrel together
with stars, broken pieces of white iron, and chippings from steel castings.
For small castings, good results are obtained with 15 to 30 min. of
tumbling.   In some cases, pieces of electrodes are added to the tum-
bling barrel for the purpose of producing a gray shining luster to the
castings.

Tumbling is the least expensive of all the mechanical-cleaning opera-
tions.   The cleaning action is one of peening that does not produce a
scrubbed surface, and many surface imperfections are not brought to
view until a machining cut is taken.

Pickling.—Steel castings are required from time to time to be cleaned
by acid-solution treatment.   This procedure is generally known as
"pickling."   The pickling of carbon and low-alloy-steel castings is an
operation not practiced so much today as in former years.   It is used
mostly today to clean sand and scale, formed in heat-treating, from deep-
cored holes or from inside passages inaccessible to pressure blasting.
Castings having passages or chambers for carrying steam or which are
subject to hydraulic flow are often required by purchasers to be pickled.
This is done so that sand or scale will not be dislodged from the casting
walls during the operation of the equipment and enter the system.

During the decade of 1920 to 1930 there was considerable use of
pickling as an inspection medium of steel castings.   Purchasers would
often, upon receipt of castings from the foundry, pickle them as a pre-
liminary operation to their acceptance.   In many cases, steel castings

were found to be cracked after being subjected to the pickling treatment. This condition led to considerable discussion between the purchaser and producer as to whether the cracks often found in pickled steel castings were caused, or merely uncovered, by the pickling operation.

Research studies carried on by the Navy Department[9] definitely proved that the surface cracking of castings generally was the result of the pickling operation. Since this time the steel foundries have stayed away from pickling wherever possible and furthermore have cautioned purchasers regarding the possible ill effects of the embrittlement of steel castings resulting from acid pickling.

Several theories have been proposed to explain the removal of scale during pickling. The one having the most popular following is that the acid dissolves the scale; however, it is common knowledge that a scale sludge accumulates in pickle tanks in the course of working. A more reasonable opinion is that the inner ferrous oxide phase layer of the scale is dissolved comparatively quickly, while the oxygen-rich phases, magnetite and ferric oxide, are practically unattacked. Thus at a scaling temperature of 1650°F., about 80 per cent of the scale is dissolved, and the remaining 20 per cent falls from the piece and accumulates as a sludge in the pickling tanks.

It is difficult to determine what occurs in the pickling of low-temperature scales; although these scales are thin, they require a much greater time to pickle than do the thicker ones formed at higher temperatures—a result that is not unexpected, since the FeO phase is largely absent from this type of scale.

The removal of sand is largely accomplished through the use of a hydrofluoric-hydrochloric acid pickle. The hydrochloric acid loosens the sand by attacking the steel matrix, and the hydrofluoric acid dissolves the sand grains.

The chief variables that affect the time of pickling are the type and concentration of the acid, the temperature of the bath, the state of agitation, the concentration of iron salts, and the type and amount of inhibitor used. The rate of pickling is directly proportional to the acid content at constant temperature, provided that the bulk of acid solution is sufficient to avoid appreciable alteration in concentration during pickling.

In general, pickling inhibitors should be used, not only for a saving in acid that is likely to be made but as an insurance against overpickling and wastage of valuable material. They are helpful in preventing acid from attacking the clean base metal of the casting, and even though they increase the time of pickling, they should be used in practice if the best procedure is desired.

The following methods of pickling employ the best commercial methods.[9] The processes should be technically controlled as to tem-

perature and chemical requirements, and routine tests must be made if the best results are to be obtained.

Hydrofluoric Acid Pickling. Use.—The primary use of hydrofluoric acid is to remove sand in the deep-cored pockets in castings.

INITIAL CONCENTRATION

	Per Cent by Volume
Hydrofluoric acid	7
Hydrochloric acid	8.5
Water	84.5

TEMPERATURE.—To obtain the best results, the temperature should be from 50 to 60°C. (122 to 140°F.) for approximately 1 hr. However, good results are obtained when the solution is at room temperature.

SANDBLASTING.—Before pickling, all loose sand should be removed; otherwise acid would be wasted in dissolving sand that could be removed more economically by sandblasting.

NEUTRALIZING BATH.—After pickling, the castings should be washed with water and then placed in a sodium carbonate bath to neutralize the acid remaining.

INHIBITORS.—As in sulphuric and hydrochloric pickling, an inhibitor should be used in order to minimize the acid attack on the steel.

EMBRITTLEMENT.—Pickling in hydrofluoric acid produces practically the same embrittling effects as in sulphuric acid of equivalent concentration.

Sulphuric Acid Pickling. INITIAL CONCENTRATION.—A concentration of from 5 to 10 per cent is used most commonly, commercially. A concentration of 50 g. per l. of H_2SO_4, 4.85 per cent, is suggested by taking one part by volume of H_2SO_4 and 26 parts by volume of water (65 lb. per 100 gal. water). Density is 1.062.

TEMPERATURE.—Sixty to 70°C. (140 to 158°F.) is the best temperature. Heat is maintained by steam coils or steam injection.

MAINTENANCE.—Keep the free-acid content up to 50 g. per l. of H_2SO_4 (4.85 per cent, density 1.062) by additions of sulphuric acid as necessary, but on no account add more than a total of 250 lb. of sulphuric acid per 100 gal. of tank volume.

FINISHING.—Additions should be discontinued when the specific gravity of the liquor reaches 1.17. When the final addition has been made, the activity of the liquor should be maintained as far as practicable by raising the temperature, and, if possible, the tank should be changed over to work as a preliminary rough pickle. When the activity has fallen below the point where the free-acid content is below 10 g. per l., the

liquor should be discarded. The discarded liquor should approach 10 g. or less of H_2SO_4, 80 g. per l. of iron, and specific gravity 1.20.

INHIBITORS.—Inhibitors that are acceptable under the Navy Department Specifications 51-I-2, Mar. 1, 1930, will furnish good results such as quinoline ethoidide 0.01 per cent, or Rodinal, etc.

IMMERSION TIME.—Immersion time is a variable depending on the size of the casting, the time and temperature of heat-treatment, and the rate of cooling.

NEUTRALIZING BATH.—The bath most commonly used is a 10 per cent sodium hydroxide solution at a temperature of 60 to 80°C. (140 to 176°F.).

REMOVAL OF ABSORBED HYDROGEN.—Steel that is rendered brittle by pickling becomes less brittle upon standing at ordinary temperatures. The recovery is never quite complete, but sections of about $\frac{3}{8}$ in. reach a maximum in about 3 days. In air at 100°C. (212°F.) the recovery takes place more rapidly, about 3 per cent of the previous time, while at 150°C. the recovery is about 0.3 per cent of the original time of 3 days. Heating in water at 100°C. (212°F.) is the best practice, and maximum recovery occurs in from 2 to 5 min. on sections up to $\frac{3}{8}$ in.

Hydrochloric Acid Pickling. INITIAL CONCENTRATION.—A concentration of about 10 per cent is most commonly used. A concentration of 100 g. per l. of HCl (9.7 per cent) is suggested by taking one part of acid by volume per two parts (approximately) of water (530 lb. per 100 gal. of water). Density is 1.045.

TEMPERATURE.—Thirty to 40°C. (86 to 104°F.) is the best temperature. Warm up initially by steam injection or by charging warm material from the previous operation. Temperature will then be maintained by the heat of reaction.

MAINTENANCE.—Keep free-acid content between 50 and 100 g. per l. of HCl by additions of acid as necessary, but on no account add more than a total of 720 lb. of acid per 100 gal. of tank volume. Additions should be discontinued when the specific gravity of the liquor reaches 1.20.

FINISHING.—The same remarks apply here as for sulphuric acid. The discarded liquors should approach 10 g. per l. of HCl or less, 120 g. per l. of iron, and specific gravity of 1.24.

INHIBITORS.—This is the same as for sulphuric acid.

NEUTRALIZING BATH.—The same remarks apply as for sulphuric acid.

REMOVAL OF ABSORBED HYDROGEN.—This is the same as for sulphuric acid.

Embrittlement Due to Pickling.—Embrittlement of cast steel was illustrated by Briggs and Gezelius[10] by studies on a 0.23 per cent carbon steel cast in standard keelblock bars, annealed at 1650°F., machined and pickled in 10 per cent H_2SO_4 at 150°F. for 1 hr. The results obtained

when the tension bars were tested in the pickling solution, without any opportunity for recovery from embrittlement, are shown in Table LXXII.

TABLE LXXII.—CAST STEEL TESTED IN 10 PER CENT H_2SO_4 PICKLE
Briggs and Gezelius[10]

Condition	Tensile str., p.s.i.	Elongation in 2 in., per cent	Reduction of area, per cent
Not pickled.................................	70,000	30.5	42.5
Pickled, no inhibitor.........................	65,600	12	15.5
Pickled, quinoline ethiodide 0.01 per cent.......	67,000	16	19
Pickled, rhodine 10 per cent..................	66,000	16.5	19.5
Electrolytically pickled, no inhibitor, cathode...	66,500	14.3	16
Electrolytically pickled, no inhibitor, anode.....	66,500	19	25

There is a considerable loss in ductility when cast steel is tested in the pickle solution. Inhibitors reduce the degree of embrittlement to a slight extent but do not avoid it.

Recovery after pickling was studied by pickling cast-steel tension specimens for 1 hr. in 10 per cent H_2SO_4, washing in cold water, and allowing to stand at room temperature, as shown in Table LXVIII.

TABLE LXXIII.—RECOVERY OF CAST STEEL FOLLOWING PICKLING
Briggs and Gezelius[10]

Condition	Time in air including testing	Tensile strength, p.s.i.	Elongation in 2 in., per cent	Reduction of area, per cent
Not pickled..............	....	70,000	30.5	42.5
Pickled.................	0 (tested in pickle)	65,500	12	15.5
Pickled.................	80 sec.	68,500	12	18
Pickled.................	8 min.	68,500	21	22
Pickled.................	1 hr.	69,000	22	25
Pickled.................	24 hr.	69,000	23	26

The recovery from embrittlement is nearly complete in 1 hr. at room temperature.

The studies presented above and those made on wrought steel indicate that pickling is responsible for an embrittling effect. This brittleness is brought about by an absorption of atomic hydrogen by the steel. Pickling does cause cracking in castings, but the actual process is not known. It is known, however, that the cracking depends upon certain things: castings of a high internal stress, and the action of the pickling solution. The greater the internal stress, the greater the danger of cracks.

There appears to be no record of castings failing to complete destruction by pickling. What is usually encountered is a series of surface cracks, which in some cases penetrate to considerable depth. It is not known how these cracks begin to form, for it is difficult to see how a reduction in ductility could cause cracking unless there were some bending movement or deformation. With the yield point measurably depressed and the casting stresses higher, the slight yielding of the casting would be sufficient for crack formation. Once a beginning is made, development of the crack is merely the result of the centralization of stresses. When these stresses are relieved, the cracking stops.

Since cracking resulting from pickling is found entirely on the surface of a casting, perhaps the surface imperfections that are present in all castings and the etch pits due to acid solution may be sufficient for excessive stress concentration and crack development.

In view of the fact that high stresses are apparently necessary in causing castings to crack during pickling, it is suggested that all steel castings be full-annealed prior to pickling. Annealed castings will show hydrogen embrittlement but not to the degree that is found in stressed castings. Also, the danger of surface cracking is considerably lessened.

Hydrogen embrittlement due to pickling is mainly temporary. The steel will usually regain almost all of its original mechanical properties if allowed to stand in air for about 3 days, or if it is placed in boiling water for 1 hr. However, it will never recover all its tensile properties. The portions that are permanently lost are probably owing to a roughening of the surface, which causes a concentration of stresses when holes and cracks are etched. Therefore, any casting that has not cracked during pickling has probably not suffered a great deal as far as mechanical properties are concerned.

CUTTING AND CHIPPING

The gates, risers, and extra pads of metal are removed from steel castings by flame cutting, abrasive cutoff wheels, cold sawing, or flogging. Chipping and flame gouging and abrasive grinding are used in some cases to remove extra pads of metal from the casting and to explore casting defects in preparation to repair by welding.

Flame Cutting.—Those gases most commonly used with oxygen for metal cutting are acetylene, propane, natural gas, manufactured gas, and hydrogen. Acetylene is the gas most commonly used for removing gates and risers from steel castings.

Oxyacetylene.—The use of the oxyacetylene cutting process for the removal of gates and risers from plain steel castings has long been a routine procedure in practically all steel foundries (Figs. 247 and 248). In determining what cutting technique to use in cutting steel castings, it

is important to know the steel analysis. The alloy and carbon contents
are factors that bear on details of the cutting technique such as the need for
preheating, flame adjustment, and blowpipe manipulation. The high-
alloy steels such as the chromium-nickel austenitic steels and the
straight chromium steels become more easily cut as the alloy content
decreases. Some of the low-alloy steels are easily cut by using the same
technique as for low-carbon steels. When the carbon content is over
0.30, the low-alloy steels usually are of the deep air-hardening type, and
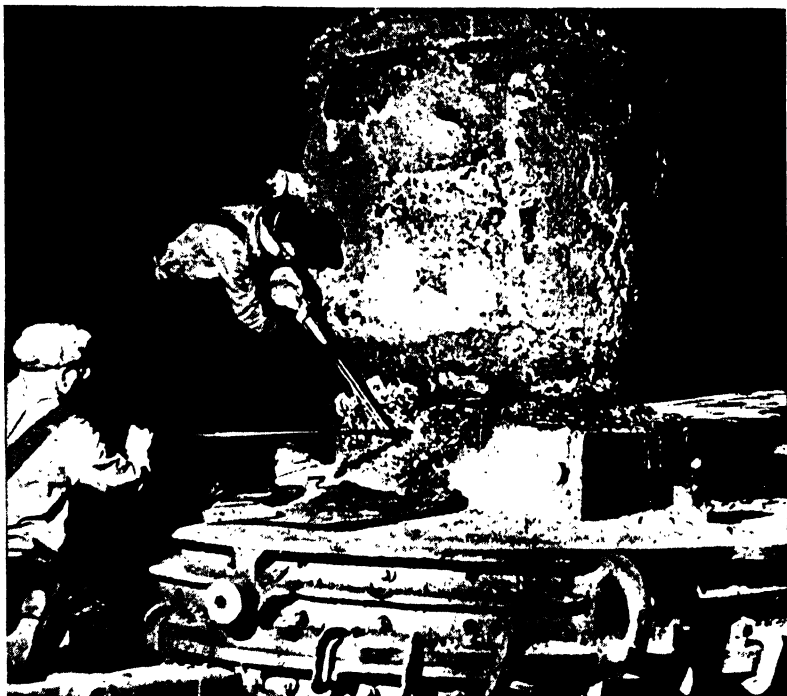
FIG. 247.—Cutting off a large riser on a steel casting.

the entire casting should be preheated to around 900 to 1000°F. before
attempting to remove the gates or risers with flame cutting.

Usually the high-tensile alloy steels with carbon contents below 0.30
per cent may be cut at room temperature without cracking, providing a
neutral preheating flame is used. For this purpose the injector-type
cutting blowpipe is used. High-alloy steels require the use of a nozzle
with six preheating orifices, whereas the nozzle for carbon steels and
low-alloy steels usually has four preheating orifices. The oxygen-cutting
and preheating pressures are approximately 30 to 40 and 6 to 10 p.s.i.,
respectively. Oxygen pressures required for cutting high-alloy castings
are about 15 to 20 per cent greater than with a similar casting of ordinary

carbon steel. Small steel castings can be placed in a combination track and casting support and flame-cut at the rate of 200 to 300 per hr., with the cut kept within an allowable tolerance of $\frac{1}{32}$ in.

Even after considerable experience in cutting metals with the oxy-acetylene blowpipe, it is still difficult for many operators to realize that oxygen cutting is a chemical reaction and the difficulty of cutting or the time required is not at all proportional to the thickness of the metal. When an operator first attempts to cut material much heavier than that to which he is accustomed, there is an almost universal tendency to use an unnecessarily high oxygen pressure for heavy cutting. This mistake, which is a very natural one to make, is probably the most frequent cause
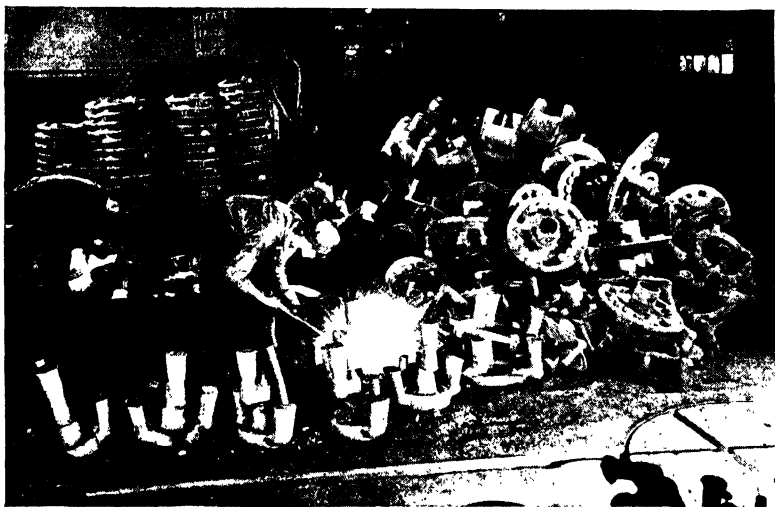
Fig. 248.—Removing gates and risers from small steel castings by flame cutting.

of trouble when first attempting to cut large risers from castings. When an oxygen pressure considerably in excess of that recommended is used in cutting thick sections, the cut after starting properly will stop altogether, and examination will show that a rounded cavity has been formed in the steel.

There is, of course, a limit to the thickness of metal that can be cut with the blowpipe alone. For the average operator the limit is probably 18 or 20 in. Steel of great thickness can be cut by means of a device known as an "oxygen lance," used either alone or in conjunction with a cutting blowpipe. Large 7- to 10-ton risers are removed by using the oxygen lance (Fig. 247).

The oxygen lance consists of a length of $\frac{1}{8}$-in. steel pipe connected with a hose to the regulated oxygen supply. The oxygen lance differs from the cutting blowpipe in that there are no heating flames to bring

the material to the kindling point.   The lancing operation may be started in many ways, but the easiest is to use the blowpipe to start the cut and to pick it up with the lance before removing the blowpipe.   Oxygen pressure is then increased by only a few pounds as the lance penetrates the metal.   The final pressure depends on the thickness of the section but is seldom more than 75 to 100 p.s.i.

Natural Gas.—Oxygen and natural gas or manufactured gas are used for cutting in a number of steel foundries, especially in areas where natural gas can be used very economically.   The only outstanding difference between the use of natural gas and acetylene in cutting action is that the preheating time required for natural gas is longer.   The tip of the blow-torch is larger in that there are usually about 10 preheating openings for the natural gas, whereas the tip for the acetylene contains either four or six gas passages.

Oxypropane.—Propane is used as a fuel for such operations as heat-treating furnaces, ladle heating, core ovens, space heating, etc.   Also, because of greater convenience and safety, it is being used for flame cutting in a few foundries.   Operators who are accustomed to acetylene have a difficult time at first to adjust the flame properly; after some experience most operators prefer to use propane, especially on heavy cuts.   The propane torch is not affected so much by sand, slag, and scale and rarely pops or backfires.   Propane is slower than acetylene in starting a cut; but this is overcome by an experienced operator, by starting on a rough spot or sharp edge.

Carbon and low-alloy steels in thickness to 4 in. are cut with about 45 p.s.i. of oxygen and 20 p.s.i. of propane.   The nozzle tip consists of a center hole for oxygen and 10 V-shaped slots for the preheating flame. For heavier sections from 4 to about 18 in., a slightly larger oxygen opening is used, and pressures of 65 p.s.i. of oxygen and 20 p.s.i. of propane are employed.

Costs of cutting different thicknesses of steel with the oxypropane torch have been obtained,[17,18] and the comparisons that have been made with the cost of oxyacetylene show lower over-all costs.   However, the foundries using propane do not report any outstanding savings.   This is probably due to the fact that they have found it advisable to use higher propane pressures than suggested by propane dealers and equipment manufacturers.

Flame Gouging.—Gouging provides a means for quickly and accurately removing a narrow strip of surface metal from steel castings.   The principle of the process is based on the fact that if a nozzle designed to deliver a relatively large jet of oxygen at low velocity is properly manip-ulated, a smooth, accurately defined groove can be cut or gouged out of the surface of the metal.   By using manipulations and nozzles of a

different kind the groove can be varied in width and depth at the will of the operator. The equipment required for gouging operations is a standard cutting blowpipe and various-size cutting nozzles. These nozzles are designated by numbers that refer to the approximate diameter of cutting oxygen orifice in hundredths of an inch.

In general, there are two techniques for gouging: one in which a groove is made progressively across a casting, as in the removal of surface cracks; the other known as "spot gouging," in which a small area is gouged out, as in the removal of shrinkage cavities or large sand spots.

For gouging on steel castings a large groove is usually desired, so that the groove will be about $\frac{1}{2}$ in. wide and $\frac{3}{8}$ to $\frac{1}{2}$ in. deep. A nozzle No. 25 or larger is used and the regulator pressure for the acetylene will be 10 p.s.i., and for the cutting oxygen it will be at least 100 p.s.i. The preheat flames are played on the spot where the cut is to be started until the surface reaches the ignition point. The cutting oxygen is then increased. In starting the gouge, the blowpipe is held so that the tips of the inner cones of the preheat flames are just touching the casting. After the cut is started, the nozzle is held so that the tips of the inner cones are $\frac{1}{4}$ to $\frac{1}{2}$ in. back of the reaction zone. The blowpipe is held so that the nozzle clears the bottom of the groove as it progresses along the cut.

The depth of the groove depends upon the size of the nozzle, the speed of progress, and the angle between the cutting-oxygen stream and the work. To cut a deep groove, the angle of the blowpipe is increased in relation to the groove. Deep cuts such as 3 or 4 in., to remove deep-seated defects, must be made by a combination of flame cutting and gouging. A vee is made in the metal by alternating use of hand cutting and gouging.

Defects that are revealed in castings by radiographic examination can be removed quickly and completely by spot gouging. The nature of the resulting cut is a narrow V-shaped groove that can be readily filled with weld metal.

This method of metal removal is much faster than that of chipping. However, as a general rule gouging should not be applied to air-hardening steels where the hardness resulting from the operation will be objectionable. No trouble will be experienced if gouging precedes final heat-treatment. Care should be given to see that gouging does not cause cracking in the higher carbon steels (above 0.30 per cent C) or in air-hardening alloy steels. Such a condition is the result of excessive localized temperature gradients as set up by flame gouging.

Scale resulting from heat-treating castings is removed by a flame-descaling method. A specially designed oxyacetylene heating head is fitted onto the cutting torch. The process is based on the principle that when a high-temperature flame is applied to the scale, or oxide skin, on a

piece of cold metal, the scale expands and breaks away from the base metal because of differential expansion between the scale and the steel.

Abrasive Cutoff Wheels.—Abrasive cutoff wheels are used in the steel foundry for removing sprues, gates, and risers. Any steel that can be sheared, or flame-cut, can usually be cut off more rapidly, often more cheaply, and always more accurately with abrasive wheels than by other methods. However, it does not follow that abrasive wheels should be substituted for other methods of cutting off on all casting jobs. If low cost is more important than high production or if accuracy and finish are of no great importance, other cutoff methods are often preferable.

The use of cutoff wheels is limited only by the size of the gates and risers. The limit of most cutoff machines is risers 2 or 3 in. in diameter, although one machine will cut off 6-in. diameter risers. One steel foundry reported a saving of 33 per cent when it switched from flame cutting to abrasive wheels. The flame so heated the castings that considerable time was required for their cooling before they could be handled for snagging. By using the cutoff wheel, the cooling time was eliminated and the amount of snagging needed was greatly reduced.

Whether the gates and risers are sheared or flame-cut, it is nearly always necessary to finish the ends by grinding. By properly selecting the cutoff wheel, it is possible in the cutoff operation to get nearly any degree of finish required, so that subsequent grinding is not needed.

Dry cutting is much faster than wet cutting and is the method most used in the steel foundry, although for cutting steel castings it has been shown by wheel manufacturers that wet cutting is the better method. In dry cutting it is desirable to use as high a speed as possible. The usual range is between 10,000 and 16,000 surface ft. per min. The time required per cut at 16,000 surface ft. per min. is one-eighth of that at 4,000 surface ft. per min., and the wheel wear is one-third of that at the lower speed. The higher speed requires about four times the power, but this extra cost is more than offset by the saving in time per cut and the increased wheel life.

Cutoff wheels are made in various thicknesses. In general it is advisable to use the $\frac{1}{8}$-in. wheel. However, deep cuts may make it necessary to use thicker wheels, for strength and for straightness of cut. The $\frac{1}{8}$-in. wheels are usually 14 or 16 in. in diameter.

For cutting steel castings, aluminum oxide is the abrasive most commonly used and 30- to 50-grit size is the most preferred. Coarse-grit, hard-grade wheels give longest wheel life in dry cutting, but they are more likely to burr the work. Because rubber-bonded wheels will stand more flexing than wheels of any other bond, they are recommended when work is fed to the wheel freehandedly as is sometimes done in cutting off gates and risers. They are strong and able to stand severe shocks and

strain better than resinoid wheels. Modern rubber wheels stand high temperatures well. The principal trouble that is likely to be encountered in cutoff operations is wheel breakage.

Chipping.—The usual procedure of miscellaneous steel castings after they have been shot-blasted and the gates and risers removed is to chip them before turning them over for snagging operations. Experience and study of chipping operations in the steel foundry have brought out the need not only of proper equipment but also of service and proper flow

Fig. 249.—Chipping operations on side-frame castings.

of work to the chipper. All chipping benches should be properly designed, standardized, and located. The trays and bins should be handy to the chipper so that time will not be lost in excessive handling of castings. Also, increased chipping efficiency comes with the segregating of castings of the same pattern.

The chipping operation consists mostly of removing fins, sharp corners, edges, and extra metal pads (Fig 249). It is preferable to remove this extra metal by chipping rather than grinding, because it results in lower cleaning costs. The removal of sharp edges and corners by grinding is always responsible for high grinding-wheel costs.

In considering chipping hammers for steel removal, it is suggested that several items be considered, such as weight of the hammer, air consumption, number of blows per minute, force of the blows, vibration, and

general versatility.   The force and number of blows per minute establish a basis for comparing the work performed by the different makes of chipping hammers; the air consumption indicates a comparison of operating costs.

Maintenance of chipping hammers is also an important feature that must be given consideration.   It has been found generally that minor repairs, such as replacing bushings and throttle levers and refitting worn cylinders with oversize pistons, can usually be made satisfactorily by the foundry.   All other repairs have to be referred to the manufacturer of the hammer.   Chipping hammers should be continually checked to see that they are in a good operating condition, and occasionally they should be tested.[16]

The type of hammer most suitable for general miscellaneous use is one that operates at about 85 lb. air pressure with a striking frequency of about 2,400 blows per min.   Such a hammer has a stroke of about 2 in. requires approximately 16 cu. ft. of air per min., and weighs approximately 12 lb.

The selection of suitable tool steels for chisel blanks should not be overlooked if operating expenses are to be kept under control.   The elimination of tools, marked and noted, that are too soft or too brittle will establish a tool specification for each specific operating condition.   It has been found generally by foundries that it is more economical to use forged-to-shape steel for standard chisels and plain forged steel for special shapes.

GRINDING

After the gates and risers have been removed, excess metal has been chipped off, and fins, brackets, and defects have been removed by chipping, the exposed surfaces are ground or "snagged" smooth to the casting contour.   Any undersirable rough surfaces are also snagged, and if necessary any extra metal pads are removed by grinding the section down to the required finishing dimensions.   The type of grinding employed is known as "rough" grinding, since the ground-surface finish is comparable to the finish obtained on the surface of the casting in the cleaned, as-cast condition.   A rough-surface finish is all that is desired, since in most cases the castings will be heat-treated following the grinding operation.   Steel foundries are not usually required to do any precision grinding.   They need only to clean the castings, which may be done on one or more of the three types of grinder: floor or bench stand, portable, or swing frame.

Floor and Bench Stands.—The most commonly used grinders in the steel foundries are the floor and bench stands (Fig. 250).   The floor stands are used for production grinding of both light- and medium-weight cast-

ings. They generally employ wheels 12 to 14 in. in diameter and are used principally on castings that can be ground while holding them against the wheel. These machines are usually equipped with two or more speed changes to permit the maintenance of the peripheral speed of the wheel as it wears and decreases in diameter. The bench stands are used for small castings and for finishing operations.

Portable Grinders.—The portable grinders are operated by a self-contained motor, by a flexible shaft, or by air. They are used widely in the steel foundry to snag off fins and remove small defects both on uninterrupted surfaces and places that are difficult to reach on complicated castings or in irregular casting passages (Fig. 251). The most common

Fig. 250.—Floor stand grinding.

sizes of straight wheels used on portables are, according to Heywood,[11] 6 by 1 by $\frac{5}{8}$ in. and 8 by 1 by $\frac{5}{8}$ in. Cone-shaped wheels are used for the grinding of small radii and inner surfaces of cored passages. The most popular size of cone wheel is $2\frac{3}{4}$ by $3\frac{1}{2}$ by $\frac{5}{8}$ in. or the same wheel in $\frac{1}{2}$-in. thickness.

Swing Grinders.—Swing-frame machines are used principally on medium and large castings with a surface that can be ground from the top side, and also for small castings from which large amounts of metal must be removed (Fig. 252). They are suspended from a trolley on a job arm or small traveling crane. It is important that swing-frame machines be designed for easy and fast handling. The wheel diameters most used are 16, 18, 20, and 24 in. Most machines have at least one speed change.

Factors of Grinding Action.—In the grinding of steel castings there are several convenient methods of modifying grinding action to meet local
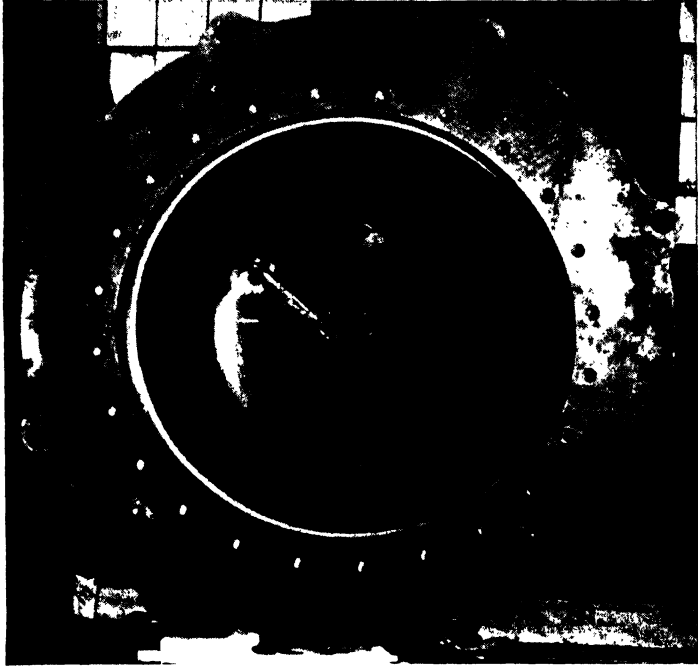
FIG. 251.—Portable grinding on an hydraulic-press casting.

FIG. 252.—Swing frame grinding.

and changing needs. The most common of these are

1. Wheel specifications, *viz.*, bond type, grain, grade, etc.
2. Wheel speed
3. Work pressure

A change in any one of these factors will directly affect the (1) rate of wheel wear, and (2) the rate of material removal. Changes in two or more of the above factors might neutralize each other.

The rates of wheel wear and material removal have a direct bearing on grinding cost, but the relative importance of each varies from plant to plant, and from time to time within a plant, depending upon the type of castings being manufactured, changing costs of labor and overhead, and, to a more limited extent, the price of wheels. It is also possible that a condition might be encountered, such as that during the Second World War, where rate of production might be of such importance that the total grinding cost would temporarily, at least, be a secondary consideration.

When these factors are all considered, it at once becomes apparent why it is not often satisfactory to recommend a definite wheel specification and set of operating conditions as equally satisfactory in all plants doing the same kind of work. A general understanding of the effects of the various changes is essential, however, in order that the variables may be chosen to the best advantage to meet changing conditions.

An understanding of the action of a grinding wheel is of great help in selecting the right wheel and in using it to the best advantage. Grinding wheels are designed to be, to a degree, self-sharpening, *i.e.*, under skilled use, as the abrasive grains dull, they will break and present new cutting edges to the work. This process is repeated until the grain has no more sharp edges, when it will be torn from its bond and a new grain will be presented. The degree to which a wheel will be self-sharpening depends upon the nature of the abrasive (the readiness with which it fractures), the hardness of the bond (the tenacity with which it holds the abrasive grains), and the stresses set up in the face of the wheel by the grinding action.

For best results the stresses should be nearly equal to the strength of both the abrasive and the bond. Then a very little increase in the stresses will fracture the abrasive grain or tear it out of the wheel. If the stresses are reduced, the bond will hold the grains after they become dull, and the wheel will glaze. If a bond permits the grains to be pulled out quite readily, then the wheel is said to act "soft"; if a bond holds the grain too long, it is acting "hard." A wheel that is too soft wears too rapidly, while the hard one results in a sacrifice of cutting action. Either raises grinding costs.

The stresses in a wheel's cutting face, because of contact between the wheel and the work, are chiefly impact, compression, and shear. It is

these stresses that tend to break the grain or tear it from the wheel. The stresses depend upon

1. The mechanical properties of the metal being ground
2. The area of contact between work and wheel
3. The ratio between wheel speed and work speed
4. The rate of traverse or cross speed
5. The rate of infeed

Items 3, 4, and 5 apply more to finish grinding than to snagging.

Hard materials resist penetration by the grain and so provide greater impact stresses than do soft materials. Tough materials increase the shearing stresses. The area of contact between wheel and work affects chiefly the compression stresses.

Wheel Specifications.—The five characteristics of a wheel in the order in which they appear in the "Grinding Wheel Marking Standard" [19] are as follows:

1. Abrasive type
2. Grain size
3. Structure
4. Bond type
5. Hardness or grade

ABRASIVE.—The abrasive used for the grinding of steel castings is aluminum oxide. Aluminum oxide is tougher than silicon carbide, the other widely used abrasive; and because it is tough, it is used as the abrasive in wheels for grinding the higher tensile-strength materials, such as carbon and alloy steels.

By altering the process of making aluminum oxide by adding chemicals to the mix it is possible to change its basic characteristics to some extent so that it can vary from a hard and tough abrasive to a very hard but fragile type. The type used for the snagging of steel castings is the heavy-duty (hard and tough) type.

GRAIN SIZE.—The best size of abrasive grain that should be employed is dictated by the material to be ground, the amount of material to be removed, and the quality of finish desired. In most cases, the larger the abrasive grain, the faster will be the rate of metal removal. Hard, brittle material will not permit the large grains to penetrate to their full depth, and therefore the advantage of the large grain is lost.

For steel castings the large abrasive grains are used when grinding with heavy pressure on a small area of contact. As the pressure decreases —as in the case of using portable grinders—smaller grain sizes are employed. The range of grain sizes varies from about 10 grit as the coarsest, to 24 or 30 grit on the portable machines, with 14 and 16 generally used on the average floor stand, although light bench stands mount

wheels of from 16 to 24 grit. The swing grinders can employ an abrasive of about 12 grit for steel castings.

STRUCTURE.—Penetration of the abrasive into the metal is easier to secure when there is comparatively wide spacing between the abrasive grains. Wide spacing assists in securing rapid material removal, since the spaces provide clearance between the grains for the comparatively large chips produced by the coarse grains. Snagging operations involving variable applications of pressure and surface operations call for wide spacing so that each grain will carry its full share of the pressure and do its full share of the work.

BOND.—Vitrified-, resinoid-, or rubber-bond wheels are used for snagging steel castings. The two types most commonly used are the resinoid and vitrified bond.

Special clays are the chief bonding materials of vitrified-bonded wheels. These wheels cost less than the resinoid wheels, but there is a definite limit to the operating speed that can be used safely. This limit is 6,500 surface ft. per min. Although vitrified-bonded wheels are strong enough for heavy work, they are subject to breakage from so-called "heat stresses" if the grinding pressure is too heavy and continuous for prolonged intervals of contact time. Only under very favorable conditions of steel-casting grinding are vitrified wheels equal to resinoid-bonded wheels when all factors are considered.

The fact that resinoid- (bonded with synthetic resin) or rubber-bonded wheels are less brittle and more resilient than vitrified wheels enables their safe use at speeds up to 9,500 surface ft. per min., and they are not subject to heat breakage. Hence, they can be made to remove stock more rapidly and will stand more abusive treatment than vitrified wheels. On steel castings where large amounts of stock removal are involved, these advantages usually more than offset the higher price.

In the more progressive steel foundries throughout the country, the change from slow-speed machines using vitrified wheels to high-speed equipment using resinoid wheels has already been made on jobs where the change would be advantageous. Some slow-speed swing frames and a few slow-speed floor stands are still in use on castings where the amount of stock removal is not great enough to justify the procurement of new high-speed equipment. Very few slow-speed portable grinders are now in use.

It should be noted that equipment designed for use with vitrified wheels may not be suitable for high-speed service, even though the speed can easily be changed. Bearing design, spindle size, weight and stability of base, and strength of protection hood must all be taken into consideration when contemplating a change to a higher speed. Amount of power

available is also an important factor if rate of production is to be increased. For these reasons it is usually advisable to obtain new equipment when a change from vitrified wheels at low speed to resinoid wheels at high speed is to be made.

HARDNESS OR GRADE.—Each type of bond can be supplied in several grades. The strength with which the bond holds the abrasive is referred to as the "grade." A soft grade increases the rate of metal removal at the expense of higher rate of wheel wear. The harder grades increase durability or wheel life at the expense of rate of production and require heavier work pressure.

It is generally true that the smaller the area of contact, the harder the wheel should be. The presence of vibration and worn parts usually necessitates the use of a harder wheel than would be required normally.

Where the grade of the wheel in use is reasonably close to the optimum for a given set of conditions, especially as regards pressure, the loss from increased wheel wear is usually greater than the gain from increased rate of cutting by using a softer grade. Changing to a softer grade in order to increase the rate of grinding must be done judiciously. If the grade in use is definitely too hard for the existing conditions, as manifested by slow cutting, need for excessive pressure, or undue heating of casting and wheel, then a change to a softer grade would be definitely advantageous and should be made.

Where it is essential that production per man be increased as much as possible, it may be advantageous to use a somewhat softer wheel, and to ease up on the grinding pressure just enough to prevent excessive wheel wear. If these adjustments are carefully made, it is likely that rate of metal removal can be considerably increased without materially affecting the total grinding cost per unit of work done. The wheel cost may be higher, but the labor cost will be less per unit of work done.

In times when rate of production is not so urgent and when labor is more plentiful, there is usually a trend toward the use of harder and harder wheels requiring greater pressure in order to make them cut. This results in reduced wheel cost per unit of work done; but unless the wage rate is materially reduced, there is seldom much change in total grinding cost.

Marking.—Each manufacturer has a code consisting of letters, numerals, or both, for indicating wheel grades. This is apt to lead to confusion on the part of the wheel operator when ordering wheels. To obviate this trouble, a simplified code has been recommended to the American Standards Association for general adoption. In December, 1944, the grading scale was revised to cover from A to Z, or from very soft (A) to extremely hard (Z). Medium is designated with symbols M, N, O. According to this grading system, the recommendations for grinding steel castings are presented in Table LXXIV.

Wheel Speed.—As indicated in the preceding paragraphs, the rate of metal removal increases with wheel speed. Theoretically, the amount of steel removed from the casting under constant-pressure grinding is the same for each revolution of the wheel at a given diameter. Hence, the rate of cut should be proportional to the surface velocity of the wheel. Within reasonable limits resinoid wheels approach a fulfillment of this mathematical prediction. Vitrified wheels are more likely to vary from it because of increased vibration at higher speed. Tests have proved that the high-speed wheel gives 2.5 to 3 times greater output for the same size of wheel.

TABLE LXXIV.—WHEEL RECOMMENDATIONS FOR GRINDING STEEL CASTINGS
Heywood[11]

	Abrasive	Grain	Grade	Bond
Carbon (0.20–0.30 per cent)				
Swing frame (low-speed)	Al_2O_3	14–16	Q	Vitrified
Swing frame (high-speed)	Al_2O_3	14	Q	Resinoid
Swing frame (high-speed)	Al_2O_3	12	Y	Rubber
Floor stand (low-speed)	Al_2O_3	14–16–20	Q	Vitrified
Floor stand (high-speed)	Al_2O_3	14–16–20	Q	Resinoid
Floor stand (high-speed)	Al_2O_3	16	Q	Rubber
Austenitic manganese				
Swing frame (low-speed)	Al_2O_3	14–16	Q	Vitrified
Swing frame (high-speed)	Al_2O_3	14	S	Resinoid
Floor stand (low-speed)	Al_2O_3	16	P	Vitrified
Floor stand (high-speed)	Al_2O_3	14–16–20	Q	Resinoid

Safety is the most important limiting factor of high wheel speed. Vitrified wheels should not be operated faster than 6,500 surface ft. per min., and for resinoid wheels the maximum speed is 9,500 surface ft. per min.

As a general rule, it is advantageous to use the highest wheel speed recommended by the wheel manufacturer for each class of wheel. The life of wheels operating at subnormal speed may be reduced as much as 40 per cent with consequent reduction in the rate of cutting. For rough grinding, such as the snagging of steel castings, wheel-speed changes are seldom made for the purpose of modifying grinding grade, as is frequently done on some precision-grinding jobs. Grinding machines should, however, be arranged so that spindle speed may be increased as the diameter of the grinding wheel is reduced through wear. A watchful eye should be kept on machines to see that they are not operating at too low a speed owing to belt slippage, insufficient power, neglect in making pulley changes when changing from one wheel to another, or failure to compensate for wheel wear.

An indication of the effect of wheel speed on the grinding cost is given in Table LXXV.

TABLE LXXV.—EFFECT OF WHEEL SPEED

	Wheel speed, surface ft. per min.		
	9,600	8,600	6,500
Wheel diameter, in..............................	16.0	14.5	11
Wheel wear (cu. in. per hr.).....................	10.5	12.2	10.8
Steel removed (lb. per hr.)......................	10.2	9.4	6.5
Ratio steel removed to wheel wear...............	0.97	0.77	0.60
Ratio wheel wear to steel removed...............	1 to 9.4	1 to 7.4	1 to 5.8
Wheel cost, cents...............................	4.4	5.5	7.1
Labor and overhead cost, cents..................	9.8	10.6	15.4
Total, cents...................................	14.2	16.1	22.5

Wheel: Resinoid, hard, coarse grit No. 12
Grinding pressure: 175 lb.
Machine: Swing frame
Work: Steel casting
Cost calculations:

Total cost per lb. of metal removed

$$= \frac{\text{No. of hr. wheel life} \times \text{labor and overhead } \$1.00 + \text{cost of wheel } \$15.38}{\text{lb. metal removed}}$$

Steel-foundry operating tests show that from a wheel speed of 9,500 surface ft. per min. down to about 8,000 ft. per min. there is no material difference in grinding expenses that the operator could not easily compensate for in the way he handled the wheel. If the wheel speed is dropped down as far as 4,000 surface ft. per min. with a high-speed resinoid wheel, it is necessary that a very light pressure be employed.

Work Pressure.—The general effect of increasing the work pressure is to increase rates of metal removal and of wheel wear in such proportions that the total grinding cost is lowered. However, at the higher pressures a marked increase in wheel wear becomes evident and increases out of all proportion to the amount of metal removed. Also, as the pressure increases, the power used to drive the wheel increases.

The other principal limiting factors to increasing wheel pressure are fatigue of the operator and danger of wheel breakage. The latter would be particularly important in connection with vitrified wheels because of increased grinding heat. If heavy pressure is employed, vitrified wheels with large-diameter openings should never be used. As pressures are increased with either type of wheel, the operators will experience increasing difficulty in controlling the grinding. Heavier work pressure usually calls for harder wheels to avoid excessive wheel wear. If pressure

is materially increased without changing to a harder grade of wheel, it is likely that the loss from increased wheel wear will cancel any advantage gained from increased rate of cut.

It is difficult to recommend specific pressures or to make any comparisons because there is no satisfactory way of actually measuring and recording the pressure being used when a wheel is in service. An indication of the effect of pressure on the item of cost is given in Tables LXXVI

TABLE LXXVI.—EFFECT OF GRINDING PRESSURE

	Grinding pressure, lb.		
	100	140	200
Wheel wear (cu. in. per hr.)...................	7.1	19.4	28.4
Steel removed (lb. per hr.)....................	5.2	12.6	21.6
Ratio steel removed to wheel wear.	0.73	0.65	0.76
Ratio wheel wear to steel removed.............	1 to 6.8	1 to 6.3	1 to 7.3
Wheel cost, cents...........................	5.8	6.5	5.6
Labor and overhead cost, cents...............	19.2	7.9	4.6
Total, cents..............................	25.0	14.4	10.2

Wheel: Resinoid, hard. Coarse grit No. 12
Wheel speeds: 9,600 surface ft. per min.
Machine: Swing frame
Work: Steel casting
Cost calculations:

Total cost per lb. of metal removed

$$= \frac{\text{No. of hr. wheel life} \times \text{labor and overhead \$1} + \text{cost of wheel \$15.38}}{\text{lb. metal removed}}$$

and LXXVII. Numerous attempts have been made to develop a foolproof method of recording pressures as a means of controlling the operator's use of a wheel. Some methods have worked exceedingly well as long as a supervisor is on hand, but they all prove inadequate for normal daily production.

A method used by Westover[12]for the preparation of grinding-wheel-performance record charts (see Fig. 253) is to tie known weights onto the handles of a swing grinder to produce a definite pressure weight. The actual pressure is measured by placing a scale under the wheel at the level at which the casting would be placed. An ammeter reading of the motor is then taken while the weighed pressure is held. The ammeter reading can then be used to indicate that the same pressure is maintained during the test-period grinding of a casting.

Judgment must be used in the applying of pressure, however, since a pressure that is permissible on one section of the grinding wheel is ruinous

if allowed on another section. For example, it is quite correct to employ a pressure of 100 lb. between the casting and the face of the grinding wheel, but 100 lb. pressure put on the corner of the same wheel shears off

TABLE LXXVII.—EFFECT OF WHEEL GRADE AND PRESSURE ON METAL REMOVED

	No. 1	No. 2	No. 3
Wheel grade	Extra hard	Regular hard	Same as No. 2
Work pressure	390 lb.	390 lb.	160 lb.
Time of test	1 hr.	1 hr.	1 hr.
Metal removed	33 lb.	100 lb.	33 lb.
Wheel loss	8 in.	24 in.	3 in.
Cost per 100 lb. metal removed	$15	$13	$7.50
Ratio wheel loss to metal removed	1 to 4.1	1 to 4.2	1 to 11.0

Labor and overhead at $1 per hr.
Wheel costs at $0.50 per lb.
Note: Approximately 200 lb. pressure is the best operating pressure for the regular hard wheel.

the corner without removing a corresponding amount of metal and raises wheel costs.

A survey of pressure-test records at one foundry showed that the grade of wheel that would be most satisfactory for their use should do
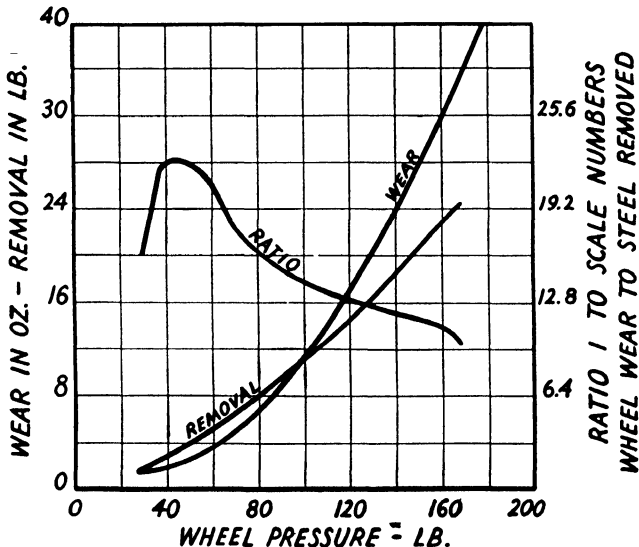
FIG. 253.—Performance record of a grinding wheel. (*Westover.*[12])

its maximum cutting at a pressure of about 100 lb. The records showed that the grinder operators could not exert more pressure than that for a

majority of the work going through the shop. It was also found that the maximum pressure for economical metal removal and speed of metal removal was different for each of the different grades of wheel tested.

Stand grinders should have large, adjustable rests incorporated into their design, with provisions for the use of pressure bars or some means other than an operator's hands and body for applying the castings to the grinding wheel. This pressure bar should be primarily an aid in keeping the casting firmly against the wheel and in eliminating any chattering. Chattering has an effect of lowering wheel life similar to wheel vibration.

Factors Affecting Grinding Costs.—In most cases grinding costs are figured on the basis of tons of castings produced. The total cost is made up of (1) wheel costs, (2) labor for actual contact, minutes, and (3) overhead for the same time.

Wheel Costs.—One of the important items in wheel costs is the proper selection and use of grinding wheels. In order to determine whether or not the proper wheel is being used, it is necessary that production tests be carried out to ascertain the following:

1. The number of pounds of metal that a wheel will remove before it is worn out.

2. The number of pounds of metal that a wheel will remove per contact hour.

In compiling this information a performance record, such as is shown in Fig. 253, should be prepared for the particular class of steel. The producer can then use this chart to make a selection of the type of wheel that will give him either a long wheel life with a comparatively slow rate of metal removal if the labor rates are low, or a shorter wheel life that is a fast metal remover if the labor rates are high. In other words, the management will be able to select a grade of wheel that will enable him to have a balanced labor and wheel cost, or a wheel cost that is either higher or lower than his grinding labor cost. To lower costs and keep them lowered, it is necessary not only to find the correct grain and grade of wheel but also to protect the wheel from unnecessary abuse.

Large gates and risers should be removed as close to the castings as possible prior to the snagging operation. It is much cheaper to cut them off with abrasive cutoff wheels or flame cutting than to grind them off with snagging wheels. Also, grinding-wheel-cost savings can be effected if castings are rough-chipped before grinding. Small, sharp areas, such as nails, vents, and fins, because of their small area, have a decided wearing effect on a grinding wheel. In addition, a casting with ends of chill nails, vents, and fins projecting from its surface is harder to handle than one with these projections removed.

To protect the grinding wheel from unnecessary abuse, it will be necessary to put an incentive on the care given the wheel by the operator.

The following are some of the things an operator can do to cause excessive wheel wear, thereby raising the grinding-wheel cost per ton of castings:

1. Exert too much pressure on the wheel.

2. Use too little pressure on the wheel, thereby allowing the wheel to bounce on the work.

3. Deliberately bounce the castings against the wheel.

4. Shear off the corners of the wheel.

5. Grind sharp corners or protruding fins of metal.

6. Grind in one spot. If a wheel is allowed to grind in one spot, it will soon generate enough heat to bring the steel under the wheel to a semi-molten state. The metal in this condition soon fills the face of the wheel; further metal removal is done by friction rather than by cutting, and steel removal is slow.

Item 1 results in excessive wheel wear and possible wheel breakage. Items 2, 3, 4, and 5 tend to put the wheel out of round and so set up vibration that may lead to wheel breakage. Frequent dressing is required when the shape of the wheel face is changed, with the resulting wheel waste.

Another item to be considered in reducing grinding-wheel costs is the difference in the cost of the different sizes of grinding wheel, if that cost is figured on the number of usable cubic inches in a wheel. The costs of usable wheels show that the smaller wheels are more costly than the larger ones. This indicates that all the grinding that can possibly be done safely should be done on the large wheels.

The condition of the equipment to which an operator is assigned is important if the amount of grinding wheel used is to be controlled. The grinder must be free of vibration caused by mechanical faults if maximum wheel wear is to be obtained. Equipment vibration is also a physical strain on the grinding operator, causing low production and excessive labor costs.

Grinding-wheel costs are related to the average weight of castings produced. The curve of Fig. 254 is drawn from the grinding-wheel-cost records of steel foundries in the Chicago-Milwaukee district.[12] The average weight of castings produced in these plants ranged from the low of 4 lb. in one plant to a high of 200 lb. in another. The curve of Fig. 254 is not intended to be an exact indicator of what grinding-wheel costs should be, but it will be useful in determining comparative costs between foundries.

Labor Costs.—Labor costs are one of the greatest causes of widely fluctuating and high grinding costs. The manner in which operators handle the grinding of castings can influence the grinding-wheel costs and the length of grinding time. For example, test studies have shown that two operators may remove practically the same amount of metal from

identical castings in the same unit of time; yet the ratio of wheel loss to metal removed will be nearly twice as great for one operator as for the other.

Tests made at a steel foundry on stand-grinder operators to determine the fluctuations in wheel costs caused by individual-operator working methods showed a varying ratio of from 1 to 3.05 to 1 to 9.44, pounds of wheel used to pounds of metal removed.

The most satisfactory labor conditions can be obtained if the grinding operator applies the correct pressure to the work and does not abuse his wheel. If wheel-saving premiums are applied without incentive-production payments, the operators will do a minimum of production work
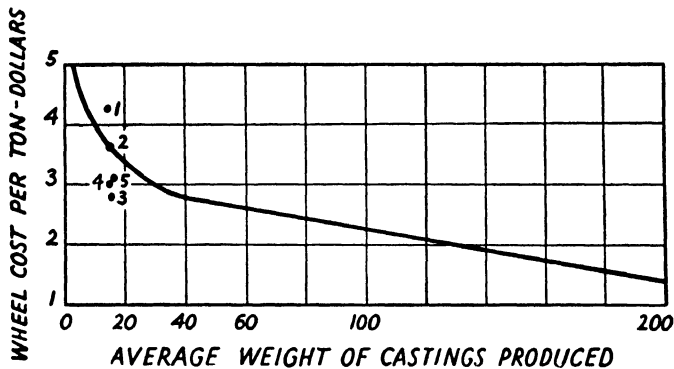
Fig. 254.—Grinding-wheel costs as related to the weight of castings produced. (*Westover.*[12])

and will depend upon the wheel-saving premiums to keep their earnings up and the minimum production per hour to keep them working the greatest number of hours.

If wage payments are made only on production, operators will have little regard for wheel care and wheel costs will be high and hence grinding costs will be higher than necessary. The best working-incentive setup for wage payments must take into consideration production and wheel saving. It is necessary that a minimum rate of production be established, which, if not exceeded by the operator, will cause him to forfeit any premium gained on wheel saving. The setting up of such a plan is described in detail by Westover.[12]

Overhead Costs.—The grinding operator does not devote his entire time to contact grinding. It is necessary for him to set up castings for grinding, shift the castings, and remove the castings after grinding. The managements of many foundries do not realize the amount of time these various manipulations take away from the contact-grinding time.

In order to reduce overhead costs to a minimum, good floors for the operator, correctly designed trays, racks, buggies, etc., for bringing the

castings to and taking them away from the grinding machines, are necessary.

Without changing the characteristics of the machine or the grinding wheel, it has been found that by merely giving the grinder efficient service and bringing the work to him in a properly designed container and enabling him to dispose of the ground castings quickly, the efficiency of the grinding unit is increased as much as 30 per cent.

Another item toward increasing the efficiency of grinding operators is to reduce the noise in the cleaning room. The loss in efficiency is the result of lessened effective effort by workmen subjected to continuous loud noise. Many steel-foundry cleaning rooms approach the high sound intensity of 100 decibels. To restore maximum effort of workmen this level must be reduced to 80 or 85 decibels, which is actually about a 50 per cent reduction in intensity. One of the best ways of doing this is to set the cleaning room off from the main part of the foundry, if possible in a special building. Also, each workman whose task—chipping or grinding—produces objectionable noise should be housed in an acoustically treated booth. The booth also can be made a part of the dust-ventilating system. Castings can be delivered to the booths and removed on roller conveyers or monorails.

Allowances for the handling of castings by the grinder operator must be made in any incentive wage plan that is prepared.

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CHAPTER XIV

HEAT-TREATMENT AND METALLOGRAPHY

STRUCTURE OF AS-CAST STEEL

When steel solidifies, crystallization proceeds from many centers or nuclei, and allotrimorphic crystals are produced that are octahedral in shape. In the case of steel solidifying in a mold, crystallization begins at the mold-cavity surface because of the chilling effect on the molten metal by the sand mold. The particles, known as "crystal nuclei," generally grow more or less uniformly in several directions towards the interior of the mold cavity, and simultaneously additional nuclei form within the metal as it reaches the solidification temperature. Eventually all the metal is transformed into the solid state by the process of nuclei formation and crystal growth.

The shape of the crystals composing the solid metal varies widely. As a rule, they are very imperfectly formed; because of neighboring crystals, however, the octahedral growth increases more rapidly in some directions than in others, and elongated crystallites, known as "dendrites," are generally obtained. Solidified cast steel is therefore made up of a great many dendrites closely interlocked. The arrangement, type, and direction of growth of these columnar dendrites are determined by the lines of heat flow peculiar to the mold being used and to the temperature gradient in the metal. An indication of dendrite formation may be evidenced by quick bleeding of any partially solidified casting. Dendrites on the solidifying face may be clearly seen, and such dendrites are illustrated in Figs. 255 and 256.

In some cases, the cooling action of the walls of the mold has a tendency to impart a columnar structure to the casting, the main axes of the dendrites being perpendicular to the mold walls. Columnar structure is particularly well shown in Fig. 257, wherein columnar dendrites exist from three cooling faces. The primary axis of the dendritic crystal forms first, followed quickly by the appearance of secondary and ternary axes. These axes grow and thicken until finally the interstices between them are filled and the dendrite is complete. Its external shape is necessarily influenced by contact with neighboring dendritic growths.

Dendrites.—The dendritic pattern of cast steel may be divided into three general regions: (1) the axes of the dendrites (primary axis and axes of the subsidiary branches), (2) the interbranchial areas (regions within a

494

single dendrite, which lie between the axes), and (3) the interdendritic areas.

As a minute crystal forms, the liquid film around it is depleted of iron. Its facets or edges are closer to a great volume of liquid than its flat sur-
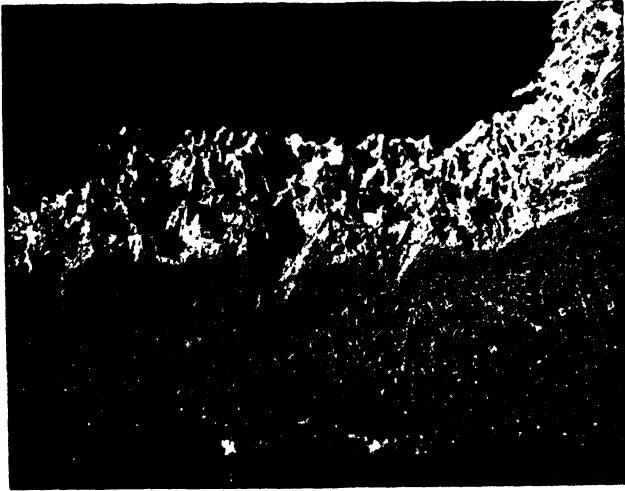
Fig. 255.—Dendritic structure as shown in a bled casting. × 1.2. (*Courtesy of Naval Research Laboratory.*)

faces, and it seems to elongate first along one of the directions connecting two opposite cube corners and then to send out branches perpendicular to the edges of the main axis of the dendrite. These branches may also
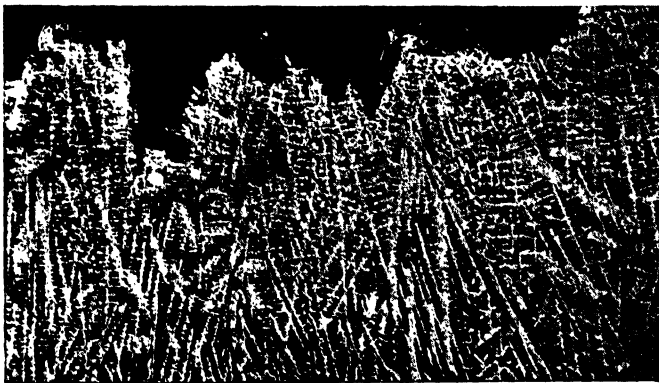
Fig. 256.—Dendritic structure showing the growing dendrites in a bled casting. × 2.4 (*Courtesy of Naval Research Laboratory.*)

branch at right angles, but the inner branches quickly lower the iron content of the remaining liquid between them. The outer branches continue to grow until they approach neighboring dendrites, after which the liquid

between the dendrites freezes only as the temperature falls through the freezing range. The dendrite in a sense reaches out its branches into the liquid in such a manner as to minimize the limitations of the rates of supply by diffusion of extra iron atoms to the crystallizing surface.[1]

In some molds, because of design, section thickness, a minimum amount of flowing metal, etc., there is a rapid formation of metal skin. Innumerable nuclei form, and there is insufficient time for appreciable diffusion or resultant dendritic segregation, since the metal is chilled so quickly through its freezing range. During the first minute the temperature gradient is very steep, and the nuclei form as small dendrites of random orientation. Very soon, with further decrease in cooling rate but

Fig. 257.—Dendritic structure in a medium-carbon steel casting. (*Courtesy of H. S. (Jerabek.*)

with a normal superheat in the metal, no more nuclei form beyond the growing dendrites; consequently, the dendrites continue growing in a columnar fashion perpendicular to the mold walls.

The typical columnar crystal zone, usually spoken of as "dendritic structure," consists of attached oriented dendrites. For the columnar structure to form, there must be some small degree of superheat in the main body of liquid metal inside the freezing zone. If, however, during the freezing the main body of remaining liquid metal approaches a zero temperature gradient, such as is found in castings of large cross section, so that a zone of adjacent liquid inside the freezing zone is at the same temperature as the inner ends of the attached columnar dendrites, then free-floating dendrites can form a little ahead of the attached columnar dendrites. The liquid layer between them is impoverished in iron, and the attached columnar dendrites cease growing. The remaining liquid then freezes by the growth of free-floating dendrites with random orientation.

The distribution of the inclusions follows the solidification pattern. The distribution of the sulphide inclusions affects the formation of the austenite grain structure and the initiation and progress of grain coarsening. The iron-manganese sulphide inclusions are interbranchially or interdendritically located, whereas in aluminum-treated cast steel the sulphide inclusions occur mostly between the dendrites.

The columnar dendritic structure in cast steel is a macrostructure and is observed best at very low magnifications, when it can be developed. In many cast-steel sections, especially carbon steels, the structure cannot be developed by etching methods. An etching treatment, to be effective, must be selective in its action; and in order that it may be selective, the sample under study must be chemically or physically heterogeneous. The fact that the columnar dendritic structure can be revealed in certain sections and compositions of steel castings by suitable etching treatments is an indication that the dendrites of steel are chemically heterogeneous, and that the dendritic segregation has not been completely eliminated by subsequent diffusion. The fact that a columnar dendritic structure cannot be revealed in cast steel at room temperature does not mean that a dendritic structure was not developed.

The view generally accepted today is that phosphorous is largely responsible for dendritic segregation. Nickel is also capable of producing intense and persistent dendritic segregation, and the addition of carbon increases this tendency.[2] Chromium intensifies the effect of nickel in producing dendritic segregation. It is believed that the effects of chromium, nickel, and molybdenum on the size and persistence of dendritic segregation depend upon their effect on the size and shape of the liquidus-solidus loop of the iron-carbon diagram and upon their effect on diffusion during and after solidification.[3]

Dendritic structure is rarely observed in the low- and medium-carbon steels.[4] Small sections, which solidify quickly, seldom show a dendritic structure on etching because of insufficient time for dendritic segregation.

The rate of cooling of the casting at the time of its solidification and the section thickness are determining factors in developing a dendritic structure that may be observed upon etching. Slow cooling, *i.e.*, a large cross section, produces dendrites that appear on etching, and a high pouring temperature will produce coarse dendrites.[5,6] Impurities such as inclusions in cast steel cause the dendrites to develop more clearly.[5]

Dendrites may be revealed in casting sections 8 to 12 in. thick, whereas in section thicknesses of less than 4 in. they may not be revealed, although the steel was the same.

It is apparent that the temperature range in which the dendrites can grow is represented by the vertical distance between the liquidus and solidus (Fig. 258). As the carbon content increases, the distance between

the liquidus and solidus increases.   The dendritic formation increases in number and size as the range of solidification is lengthened.   Thus it is easier to reveal dendritic structure in the higher carbon alloys.   Alloy additions such as chromium, molybdenum, and nickel change the shape of the solidus and increase the zone of primary crystallization and hence assist in a persistent dendritic formation that is easily revealed.

Granulation of Dendrites.—It has been established above that when solidification is complete, all cast steels are composed of closely inter-locked dendrites.   When cast steel cools from the solidus to the zone of secondary crystallization, the dendrites and the large grains constituted by the boundaries of those dendrites undergo a process of granulation. In cooling through this zone of granulation (Fig. 258), the dendrites break
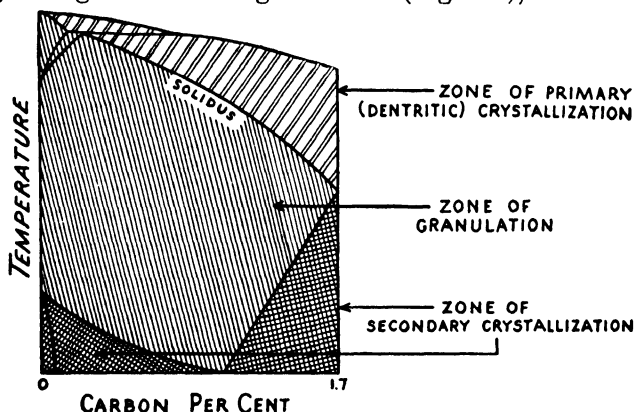
Fig. 258.—Steel portion of iron-carbon diagram.

up into austenite grains, each one with its own orientation and therefore independent crystalline existence.   The formation of the austenitic grains results in the disappearance of the dendrite as a crystalline unit.   If the dendrites are destroyed, how is it that in some instances a columnar dendritic structure can be revealed at room temperature?

The answer is that the possibility of revealing a dendritic structure through etching treatment does not depend upon the existence of den-drites but upon the existence of dendritic segregation; and that while granulation obliterates the dendrites, it does not affect dendritic segrega-tion.   The segregating element occupies the same portions of the original columnar dendrite even after the latter has undergone granulation.

It has been observed[7] that in dendritic steels not the whole dendrites but their branches are the units in which the secondary crystallization and the further transformations on reheating take place, independently of the other branches of the same dendrite, and they partly indicate that the character of the primary crystals has only subordinate importance with respect to the manufacture of commercial steel castings.

The austenite grain size in some cast steel is coarse, as shown in Fig. 259. The number of grains and their size depend upon the size of the original dendrites and the rate of cooling through the granulation zone. The grain size increases with the size of the casting and with decreased rates of cooling. In studying sections up to 15 in., it was found[9] that the primary grain size increased rapidly at first with increase of section, then slowed down until somewhere between 6 and 10 in. a maximum size

Fig. 259.—Austenite grains outlined by ferrite (as-cast structure). × 500.

seemed to be reached. With still larger sections the grain size at the center may actually decrease. This is explained by the fact that primary grain size is a function of the rate of solidification rather than the total time required for a section to freeze. Because of the latent heat of fusion, the rate of solidification is controlled by the rate at which heat can be dissipated at the surface.

According to the equilibrium diagram of iron-carbon alloys (Fig. 258), steels containing up to about 0.50 per cent carbon begin to solidify as delta iron. In plain iron-carbon alloys and low-alloy steels the significance of the delta transformation is not certain. This is largley due to

the occurrence of the peritectic reaction (delta iron plus liquid → austenite plus liquid) for carbon steels containing approximately 0.15 to 0.50 per cent carbon. Carbon cast steels below approximately 0.15 per cent carbon transform to delta iron or delta iron plus austenite. There are but few carbon and low-alloy cast steels produced with carbon contents below 0.15 per cent, and thus a true delta structure may be expected to form in cast steels for commercial steel castings. As has been stated, however, the light cores of the dendritic pattern are presumably delta iron containing very little carbon.
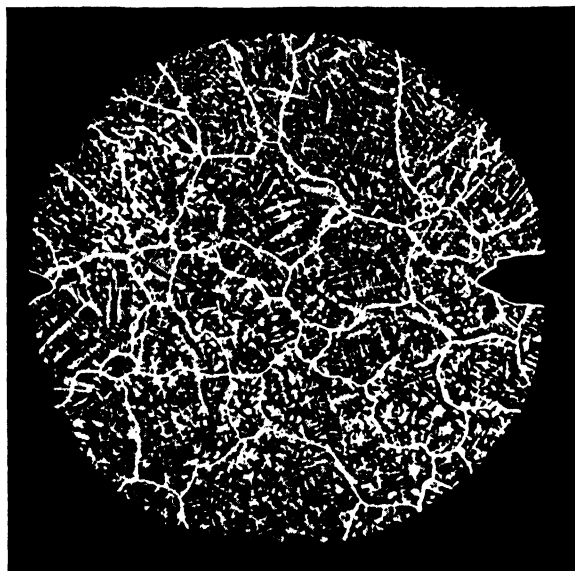
Fig. 260.—Austenite grains outlined by ferrite and superimposed on the dendritic structure. Carbon 0.35 per cent, manganese 1.40 per cent. × 5. (*Courtesy of H. A. Schwartz.*)

It is not possible to etch a sample of as-cast steel so that both the austenitic-grain size and the dendritic structure can be seen at the same time. Figure 260, showing both these structures, was prepared by first etching the specimen and manually tracing the ferrite network outlining the austenitic-grain boundaries on a film. After repolishing the specimen, it was etched to bring out the dendritic structure. The austenitic-grain tracing was placed on the negative of the dendritic structure in the printing process. It may be noted that the grains so outlined do not form with any relation to the primary axes of the dendrites; in fact, dendrites may be seen crossing several grain boundaries.

Secondary Crystallization.—A secondary recrystallization takes place when cast steel cools through its thermal critical range. If the steel is of eutectoid composition, each grain of austenite is converted into a grain

of pearlite at the single critical point of that steel. If the steel is hypoeutectoid, a certain amount of ferrite separates from the solid austenitic solution upon entering the critical range, until the remaining austenite, finally of eutectoid composition, is transformed into pearlite at the bottom of the range, at the Ar_1 point. If the steel is hypereutectoid, it is cementite, which separates in like fashion when cooling through the range. Since most cast steels for steel castings are hypoeutectoid, the steel, as it emerges from the secondary crystallization zone, is transformed into an aggregate of ferrite and pearlite presenting (1) a network structure, or
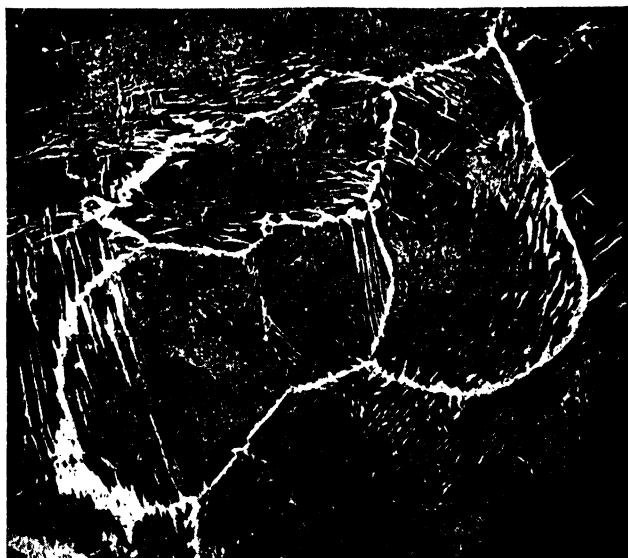
Fig. 261.—As-cast structure of ferrite network outlining the austenite grains. $\times$ 40.
(*Courtesy of Naval Research Laboratory.*)

(2) a Widmanstätten structure, according to the conditions of cooling through the critical range, the carbon content of the steel, and the section thickness.

When austenite cools through the transformation range, the proeutectoid ferrite is either rejected to the grain boundaries, giving rise to the network structure (also described as cellular[10] and globular[7] as shown in Figs. 261, 262, and 263), or precipitated in part along crystal (octahedral) planes within the grains, giving rise to the Widmanstätten structure as shown in Figs. 264, 265, and 266. If the cooling through the transformation range is too rapid to permit complete ferrite separation, a network structure is produced. A Widmanstätten structure indicates a complete separation of the ferrite and results from slow cooling through the transformation range. A large grain size in the original austenite is favorable to the development of the Widmanstätten figure, apparently
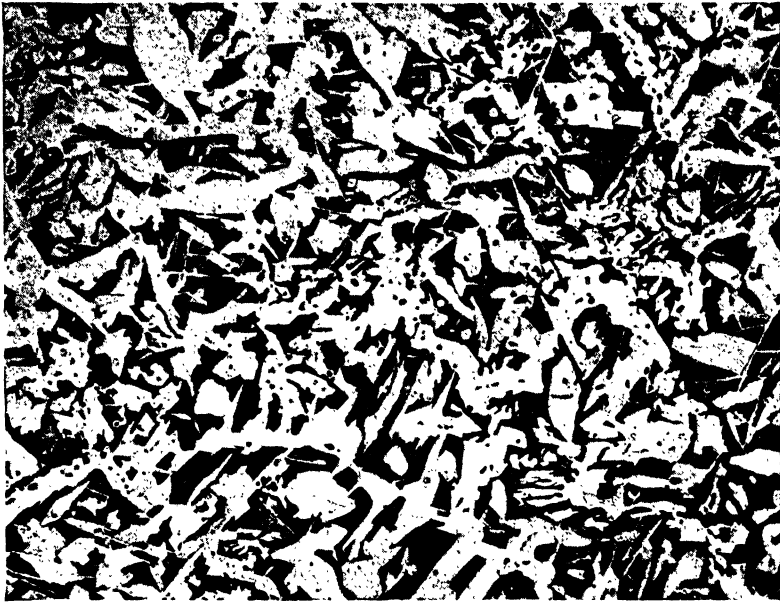
FIG. 262.—As-cast structure. Ferrite network. Carbon 0.24 per cent. × 200.
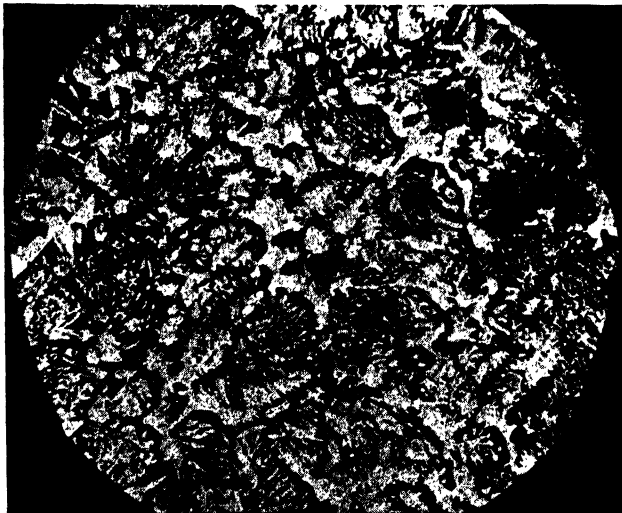
FIG. 263.—As-cast structure. Ferrite network. Carbon 0.27 per cent; 3-in. section. × 25.

because there is less opportunity in the larger grains for diffusion to and from the grain boundaries.[11] It has been reported that in hypoeutectoid alloys, ferrite forms in plates on the (111) planes of the austenite, and that the orientation of the ferrite of the Widmanstätten structure is similar to that in martensite needles.[12]

Fig. 264.—Widmanstätten structure, medium-carbon cast steel.   × 40.   (*Courtesy of Naval Research Laboratory.*)

Fig. 265.—Widmanstätten structure with ferrite at austenitic grain boundaries.   × 50.

It has been assumed that phosphorus plays an important role in the secondary crystallization of cast steel.[4]   Recent experiments[7] reveal the verity of the assumption that before the rejection of ferrite from austenite grains, the originally uniform distribution of the carbon first becomes nonuniform by the migration of the carbon away from the phos-

phorus-rich parts of the austenite grains toward their center. The first ferrite crystals that form at the beginning of the secondary crystallization appear to surround inclusions as though those inclusions acted as

FIG. 266.—Widmanstätten structure, 0.34 per cent carbon steel. × 200.
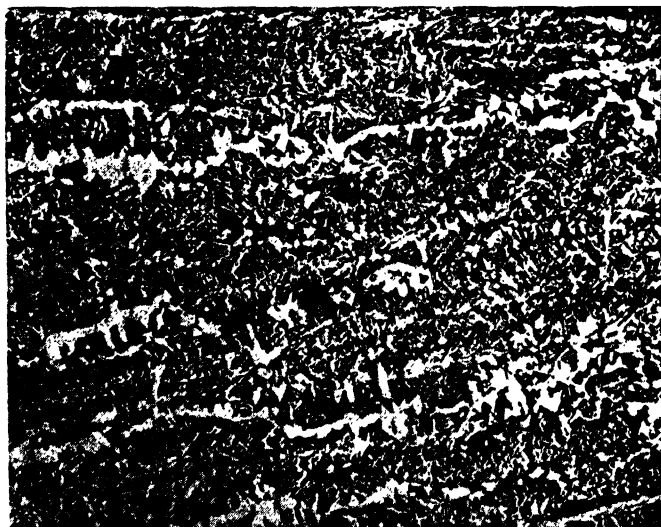
FIG. 267.—As-cast structure of ferrite network revealing a dendritic pattern. × 80.
(*Courtesy of Naval Research Laboratory.*)

nucleuses for the ferrite formation. However, the enrichment of phosphorus causing carbon migration resulted in their formation at these points.

The secondary structure, consisting of a ferrite network, may often reveal a dendriticlike structure (Fig. 267) because of dendritic segrega-

tion, particularly of phosphorus, and the formation of the ferrite network in the phosphorus-enriched areas of the primary dendritic structure. This secondary structure, as a function of the carbon content and wall thickness, is discussed in the literature[1] and is illustrated[1] in Fig. 268. The hatched area in the figure includes the compositions and cross sections in which ferrite is present—partially, at least—in the form of a Widmanstätten structure. Outside this zone the ferrite and the pearlite are arranged according to the primary structure, ferrite network, which exhibits the pearlite in globules.
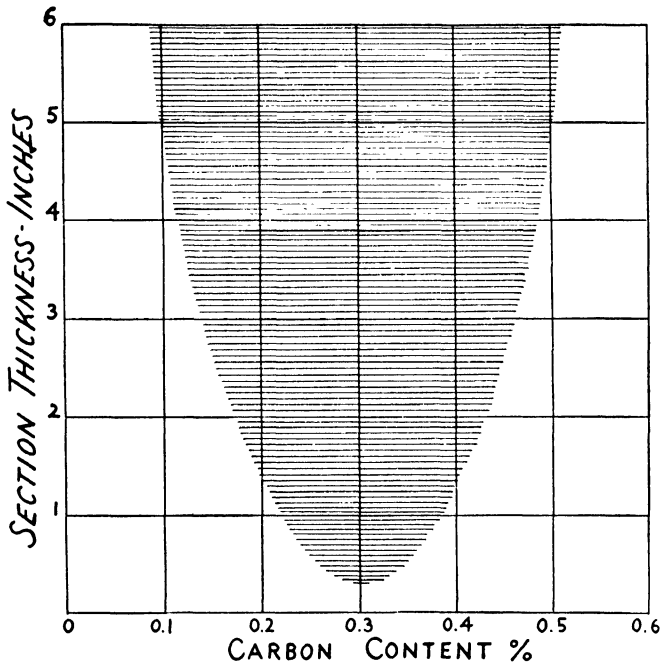
FIG. 268.—Secondary structure as a function of the carbon content and wall thickness. (*Vero.*[7])

If cast steel were observed in the as-cast condition at atmospheric temperature conditions on polished and etched sections, the following structures could be noted:

1. Dendritic. A macroscopic structure similar to Figs. 255, 256, 257 and may be observed.

2. Primary austenitic grains, as a microscopic structure. These grains are usually outlined by a ferritic network such as shown in Figs. 259, 260, and 265 or by chains of inclusions.

3. A Widmanstätten structure (microscopic), as shown in Figs. 264, 265, and 266. In Fig. 265 the primary austenite grain is also shown by a ferrite network.

4. A network, a cellular, or a globular structure (microscopic) such as shown in Figs. 261, 262, 263.

If the cross section of a casting were examined macroscopically in the as-cast condition on a polished and etched surface, it is probable that the structure would correspond to one of the following:

1. A casting cross section consisting of three zones, generally: a fine-grained rim zone, the columnar dendrites, and, finally, the inner part consisting of equiaxed crystals. The last named zone may also reveal a dendritic macrostructure. The depth of the columnar crystal zone depends upon the degree of superheat of the steel at the time it is poured. As the temperature rises, the zone becomes less distinct until it fades out entirely.[7]

2. A completely columnar dendritic cross section.

3. A completely equiaxed crystal structure with no columnar dendritic indications. The microscopic structure may be (1) coarse granular, network, or globular or (2) formed as plates (Widmanstätten).

STRUCTURES OBTAINED BY HEAT-TREATMENT

The Widmanstätten structure and the large network-type of structure are responsible for low ductility and impact properties in cast steels above approximately 0.20 per cent carbon; and one of the primary reasons for the universal annealing or normalizing of steel castings is to eliminate this structure. In hypereutectoid steels the Widmanstätten plates are composed of brittle cementite so that in higher carbon steels it is even more essential to avoid this structure. The as-cast structure has only an indirect influence on the response of cast steel to subsequent heat-treatments, and this influence is principally one of its relation to segregation. A uniform fine-grain structure can be obtained in steel castings; but to obtain it, a preliminary heat-treatment must be given the castings in order to produce structures as homogeneous as possible within a reasonable time. Some grain refinement is achieved during this preliminary treatment; but where a fine, uniform grain is desired, a second or even a third treatment is usually required.

Homogenization.—If cast steel is heated for a short time at reheating temperatures slightly above the upper critical temperature, only a refinement of the grain size is obtained, while the distribution of the ferrite and pearlite is nearly the same as in the cast state. Usually there is a conglomerate mixture of grain sizes, some very large, others quite small. The irregularity in size is apparently caused by dendritic segregation that has not been satisfactorily eliminated by diffusion during heat-treatment.

The reheating of a cast steel causes a change in (1) its grain size and (2) arrangement of the structural elements. The latter point is impor-

tant since a nonuniform distribution of the constituents *i.e.*, ferritic areas) is harmful to the properties of the steel. A uniform structure as obtained by homogenization improves the yield-point values and the impact properties. This improvement is particularly noticeable in castings of large cross sections. Improper reheating temperatures and times produce nonhomogeneous austenite. From nonhomogeneous austenite only structures showing a distribution of ferrite and pearlite
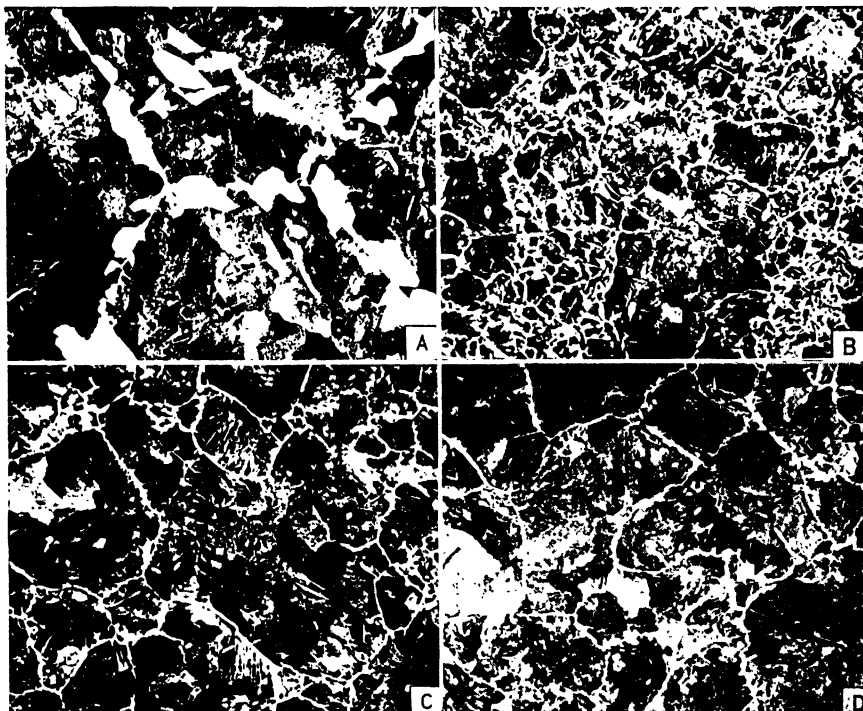
Fig. 269.—Effect of preliminary heat-treatment of medium manganese steel castings, cast and heat-treated as 1-in. sections. × 75. (*Sims.*[13]) (*a*) As-cast structure; (*b*) Air-cooled after 1 hr. at 1550°F.; (*c*) Air-cooled after 1 hr. at 1650°F.; (*d*) Air-cooled after 1hr. at 1750°F.

according to the primary structure may be obtained, even by rapid cooling. Slow cooling produces very clearly defined dendritic or network secondary structures.

The first requirement for securing a structure in which the arrangement of the constituents differs from that in the as-cast structure is the production of a homogeneous austenite. The temperature of heating is the primary consideration in homogenization. The effect of heat-treating temperature on the structure of cast steel has been studied by Sims[13] and is illustrated in Fig. 269. A medium-manganese cast steel (carbon, 0.30 to 0.40 per cent; manganese, 1.25 to 1.50 per cent) was heated at

various temperatures and air-cooled.   The 1550°F. temperature was too low to obtain proper diffusion with accompanying uniformity of grain size.   It was necessary to increase the heating time considerably in order to obtain homogenization.   In such an instance the grain size is too large, since the small grains do not grow but are absorbed by the larger ones. If the temperature is increased to 1650°F., a fairly uniform large-grain structure is obtained, and mechanical tests show that homogenization is practically complete.   If a much shorter heating time at this tempera-ture is employed, a structure similar to that of the 1550°F. treatment is obtained.   At a higher temperature the structure developed at 1650°F. can be obtained in half the time.   From this it can be concluded that
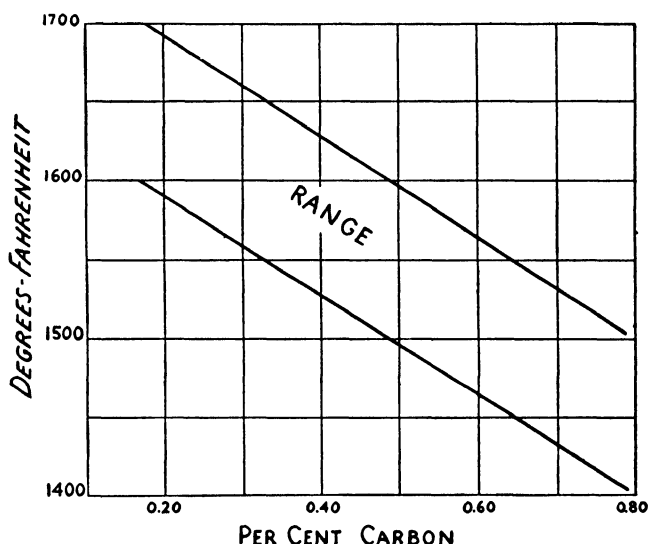
FIG. 270.—Temperature range for the homogenization and heat-treatment of carbon steel castings.   (*Metals Handbook.*[14])

starting from a point definitely above the upper critical temperature (A₃), each 100°F. increase in temperature apparently doubles the initial rate of diffusion and grain growth.   This relation does not hold for long periods, since diffusion slows up as uniformity is approached. Grain growth is rapid at first at any given temperature and then slows down to virtual cessation; thus as the temperature rises, the final average grain size increases.   The rate of diffusion, therefore, increases rapidly with increasing temperature; and in order to save time in the commercial manufacture of steel castings, a temperature somewhat above the upper critical range is used.   However, since both the rate and extent of aus-tenitic grain growth increase with temperature, too high a temperature is not desired.   A compromise temperature that is seldom more than 200°F. or less than 100°F above A₃—the upper critical temperature—is

used for carbon cast steel.[14] Temperatures in this range give diffusion in a reasonable time without excessive grain growth (Fig. 270).

Since the degree of segration is increased by the slower freezing of heavy sections, the time required for diffusion varies with the section of the casting. Castings with very light sections can be homogenized in 1 hr. or less. For heavier sections a rough guide is used that is generally 1 hr. per in. of heaviest section. This rule requires some modification. While 1 hr. is usually enough for a 1-in. section, the size of the grains and the degree of segregation do not vary as a linear function of the section but apparently reach a maximum somewhere between 9 and 12 in. It is
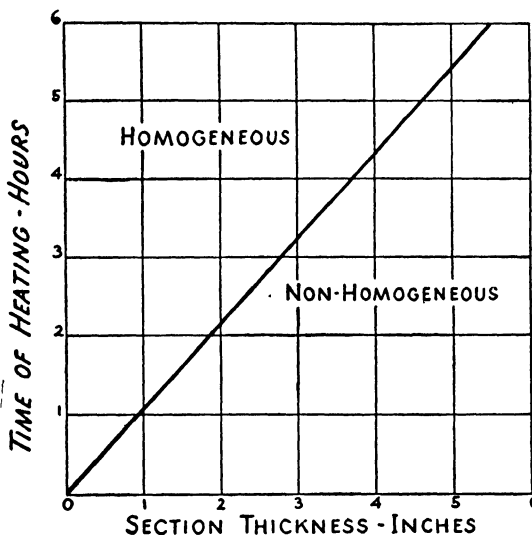
Fig. 271.—Conditions for obtaining homogeneous austenite on reheating cast steel. (*Vero.*[7])

doubtful, therefore, whether or not any casting is benefited by homogenizing it more than 10 to 12 hr. This may not apply to certain massive castings weighing upward of 100,000 lb. each.

Using carbon steel sections below 6 in. some studies have been made on the conditions for obtaining homogeneous austenite upon reheating.[7] The chart of Fig. 271 shows the amount of time necessary for 0.25 per cent and 0.51 per cent carbon cast steels to attain homogeneous conditions at 1580 and 1508°F., respectively. These temperatures are 90°F. (50°C.) above the A_3 temperature. The times necessary for homogenization increase in a manner that is approximately linear with the wall thickness. The values shown in Fig. 271 indicate that slightly longer times are required than proposed by the general rule of "1 hr. per in." However, if a higher temperature were used, such as 1650°F. for the 0.25 per

cent carbon steel, the general relationship of 1 hr. heating per in. of maximum section would be approached.

The homogenization temperature and time at temperature for cast alloy steels ordinarily depend upon the alloy content. For the usual low-single-alloy steels, such as medium-manganese steels and steels containing low percentages of nickel, copper, vanadium, and chromium, the homogenization temperature range is that as given in Fig. 270 of from 1550 to 1700°F., depending upon the carbon content. Some of the steels containing 0.50 per cent molybdenum or greater usually require higher homogenization temperatures such as 1750°F., and alloy steels of combinations of nickel, chromium, and molybdenum often require still higher homogenization temperatures. Since these alloys play a prominent part in assisting dendritic segregation, it is frequently necessary to employ high homogenization temperatures in order to obtain certain maximum special properties such as are needed for ballistic requirements. Temperatures of 1800 to 2050°F. for rather long periods of time are found necessary in obtaining complete homogenization. For example, a steel containing approximately 3 per cent nickel, 1.50 per cent chromium, and 0.75 per cent molybdenum may require a homogenization temperature of 2000°F. for 10 hr. for a 2-in. section.

Not too much correlated information is available on homogenization of alloy steels. Acceptable mechanical properties are obtained for low-alloy steels by following the practice of homogenizing 1 hr. per in. of heaviest section at temperature of 1550 to 1700°F. This practice is the usual commercial practice. Apparently such homogenization is not sufficient for the ultimate development of special properties for ballistic requirements, and both times and temperatures are greater. The scattered information at hand indicates that homogenization time cannot be based upon so many hours per inches of section. A 3-in. section may require 12 hr. at 1950°F., whereas a 2-in. section may require 10 hr. at the same temperature.

Ballistic values are increased by long homogenization times. For example, two 2½-in. sections of the same alloy cast steel were homogenized at 1950°F. One plate was homogenized for 10 hr. and the other for 20 hr.; both plates were quenched and tempered. The 20-hr.-homogenized plate had 10 per cent increased ballistic values over the 10-hr.-homogenized plate, which also had acceptable ballistic values. The former also had an increased tensile strength and Izod impact value. The time that a section of alloy cast steel is maintained at the homogenization temperature is different for the development of maximum ballistic values from what it is for normal mechanical property requirements.

If ballistic requirements are met with a homogenization of 2000°F. for 8 hr., then at a temperature of 1800°F., a time of 16 hr. is required,

while a considerably longer time of 36 hr. is necessary at 1700°F. It may be concluded, therefore, that the greater the casting section, the slower the rate of solidification in the mold, with the result that there is a greater degree of dendritic segregation. This segregation is also accentuated by the addition of certain alloys. The time required for diffusion—essential for the establishment of homogenization—increases as the segregation becomes greater.

High-temperature or extra-long homogenization periods are not necessary for the improvement of the tensile properties of cast steel. The employment of the high-temperature homogenization is not, as a rule, reflected in an improvement of these properties. Long soaking times at high temperatures were at one time considered necessary until material to the contrary was published.[8] In the heat-treatment of heavy sections, it has been demonstrated[15] that long soaking times at temperatures just above the critical are not conducive to the best properties in large low-carbon cast-steel sections. It has been shown[16] that no decided improvement is obtained in properties by air-cooling low-alloy steels in sections up to $1\frac{1}{2}$ in. from 1800°F.

The entire story of time and temperature for homogenization is not known. Most studies have been made on coarse-grained silicon-killed steels, and more consideration need be given to the finer grained as-cast structure of aluminum-killed steels. Probably the most important field for study is the rate of cooling following homogenization. Timmons[16] shows that cooling rates have a pronounced effect on segregation. The cooling rate must be increased as the section increases, and the composition must be controlled, to obtain a uniform section structure throughout.

After the homogenization treatment, which may be either furnace cooling (annealing) or air cooling (normalizing), a second heat-treatment may be given for the purpose of producing a uniformly fine-grained structure and hence better mechanical properties. If the homogenization treatment has been properly carried out, the second treatment becomes a matter of recrystallization to obtain a uniform fine-grain structure by forming as many crystals as possible and by preventing the growth of some at the expense of others. Since grain growth is a function of temperature and time, a fine grain can best be obtained by heating for a short time at a temperature just above the critical range, as shown in Fig. 270. Repeated heating and cooling of steel castings through the critical range tend to further refinement of the grain. Double and triple heat-treatments are employed in order to meet special property requirements.

Stress-relief Heat-treatment.—Stress-relief heat-treatment involves the heating of castings to a temperature usually below the critical point,

such as in the range of 1000 to 1200°F.   The annealing heat-treatment
may also constitute a stress-relief heat-treatment.   Figure 272 shows the
influence of temperature and time on relieving stresses in a medium-
manganese steel that had an initial stress of 40,000 p.s.i.[17]   It is appar-
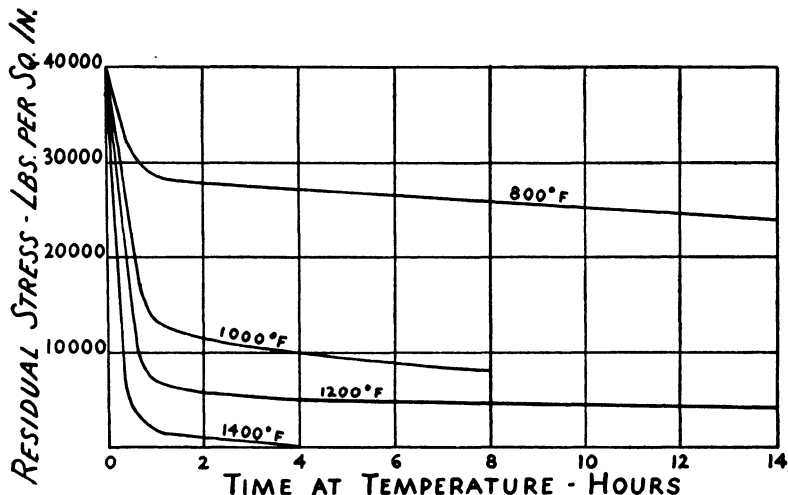
Fig. 272.—Influence of temperature and time on relieving stress introduced in manga-
nese steel (0.21 per cent carbon, 1.44 per cent Mn) by reason of straining beyond the yield
point.   (*Hiemke.*[17])
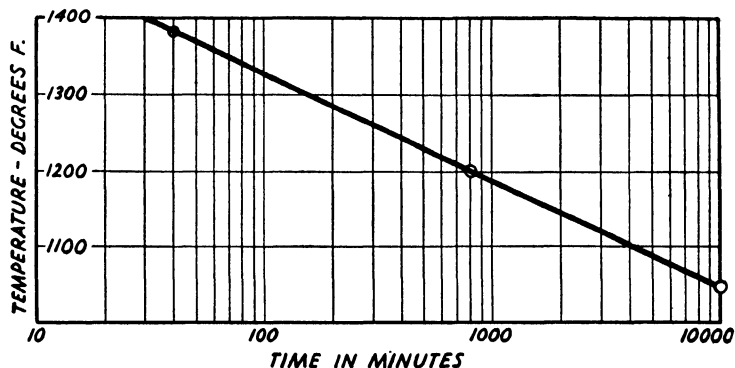
Fig. 273.—Time-temperature relation for 100 per cent stress relief for carbon steel castings.
(*Rominski and Taylor.*[19])

ent that the temperature at which stress relieving is carried on is of
more influence than the length of time at that temperature.   It is also
reported[18] that for mild carbon steel (carbon, 0.20 to 0.35 per cent), 98
per cent stress relief can be obtained in 1 hr. at 1200°F.

It is possible to determine the time and temperature required for
release of stress for carbon cast steel[19] from the chart of Fig. 273.   In

practice the time required at any given temperature for complete stress relief is less than that shown by this chart, if the castings are heated with the furnace.

There is no information as to the character of the stresses set up on the rate of subsequent cooling of the steel to room temperature. All that is known is that if the castings are cooled slowly and uniformly to about 800°F. and then air-cooled, no troubles will be encountered because of distortion during machining and in service.

Stress relieving below the critical range does not cause grain refinement, and the treatment is therefore seldom used alone. It is used extensively after heat-treatment above the critical range; or after performing an operation which may result in the setting up of stresses in castings such as in welding, scarfing, or burning; or in machining.

The procedure for tempering after accelerated cooling of castings is similar to that of stress relieving. The tempering operation is also a stress-relieving treatment.

Annealing Heat-treatment and Structure.—The annealing heat-treatment is employed for the following reasons:

1. To produce an allotropic change that causes the steel to recrystallize, thus breaking up the very large crystals formed during solidification.

2. To homogenize, or correct by diffusion, the segregation of constituents, known as "dendritic segregation," which occurs during the solidification of steel.

3. To release casting stresses and prevent the formation of stresses caused by differential cooling.

4. To improve machinability of the steel.

The annealing treatment consists of uniformly heating the casting to within the range shown in Fig. 270, if it is a carbon steel casting, for the approximate time of 1 hr. per in. of maximum cross section of the casting, and slowly cooling the casting in the furnace to temperatures approaching room temperature. For low-alloy-steel castings a heating-temperature range of not less than 100 and usually not more than 300°F. above the upper critical temperature is used. The length of the homogenization period depends upon the type of alloy steel being treated.

The recrystallization is completed as soon as the critical temperature range is passed; but since the rate of diffusion is rather slow at this temperature, a higher temperature is chosen. The holding time at temperature is the same as outlined under the section on homogenization. At the end of the holding period the heat is cut off and the furnace allowed to cool with furnace doors and ports closed. When the temperature of the furnace reaches a low temperature such as, for example, 750°F., the furnace is opened and the castings are removed and allowed to cool to room temperature in air.

The structure of annealed cast steel (0.27 per cent C) is shown[8] in Figs. 274 and 275. The effect of annealing is to refine the grain by recrystallization; but the dendritic pattern, as shown by the ferrite network, still persists on a macroscopic scale for low- and medium-carbon
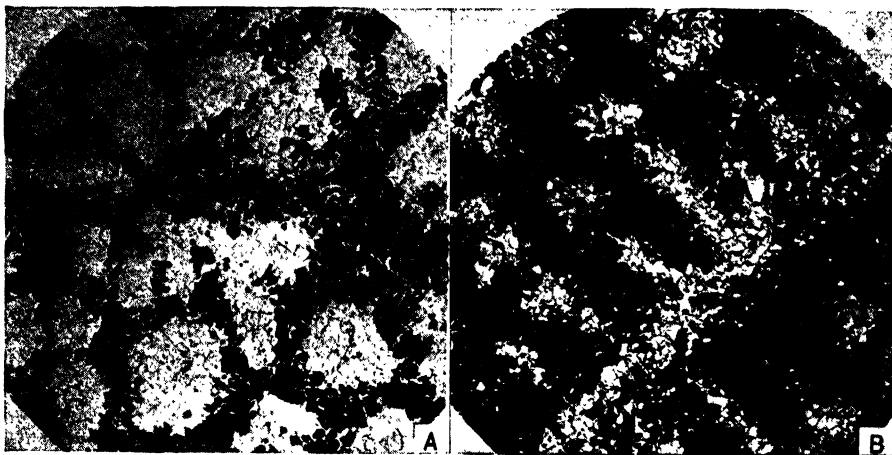
Fig. 274.—(*a*) Annealed 0.25 per cent carbon cast steel, 1650°F., × 75; (*b*) annealed nickel-manganese cast steel, 1700°F, carbon 0.33 per cent, nickel 1.60 per cent, manganese 1.10 per cent, × 75.
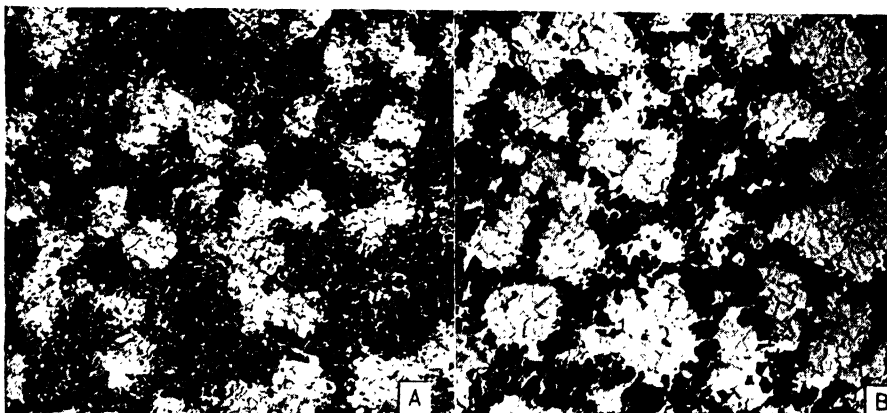
Fig. 275.—(*a*) Annealed 0.33 per cent carbon steel, 1650°F., × 75; (*b*) annealed manganese-nickel-molybdenum cast steel, 1650°F., carbon 0.32 per cent, manganese 1.20 per cent, nickel 0.60 per cent, molybdenum 0.25 per cent, × 75.

steels, and recrystallization has taken place *in situ* or has been dominated by the dendritic pattern. What appear to be islands of ferrite are really thick ferrite rims at the grain boundaries of the original austenite grains. Instead of a thin ferrite network, the thicker ferrite boundaries exist because of the heavy sections and the slow cooling of the castings (cooling

rate of approximately 0.5°F. per min. through critical range). If the cooling rate is faster, the structure will appear more normal, as shown in Fig. 276.

The section thickness affects the appearance of the annealed structure. As the section mass increases, the cooling time is increased and the ferrite islands increase in size, as shown in Fig. 277. The ferrite grains also increase in size, rapidly at first with increase of section and then slowing down until, somewhere between 8 and 12 in., a maximum size seems to be reached.

It was stated that the dendritic pattern dominates the position of formation of the ferrite and pearlite after reheating. In fact, in some cases the dendrite structure can be revealed macroscopically after reheating, its appearance being similar to that found in the as-cast condition. This so-called "dendritic" structure is not a true one; it is rather pseudodendritic—an imprint of dendrites that have disappeared.[9] The structure shows up on acid etching or heat tinting, due in part to slight differences in composition, but mainly to differences in density between the parts composing the pattern. Tests tend to show that the appearance of the dendritic pattern in the annealed

Fig. 276.—Annealed carbon cast steel. Ferrite network. × 50.

state does not produce unfavorable results. Impact specimens taken parallel to the dendrites and compared with specimens taken across the dendrites show no difference.

A comparison of the properties obtained in the as-cast and annealed conditions of carbon steels is shown in Table LXXVIII. A full review of annealing heat-treatments for carbon and alloy cast steels and corresponding properties is given in the "Steel Castings Handbook."[20]

The structure developed on annealing gives the best distribution of the constituents for general machining, provided the analysis of the steel is such that this structure is not too soft or too hard. Coarse laminar pearlite gives the best machinability, in the hardness range of about 130 to 210 Brinell, this hardness range covering the majority of cast-steel analyses produced today. Carbon steels under 0.20 per cent carbon should be machined in the as-cast condition, since the annealed structure is too soft for good, all-round machining qualities. If the hardness of the steel as annealed is over 210 Brinell, machinability can be increased by

spheroidizing the carbides through heating the steel for a long period of time at a temperature slightly below its critical temperature.
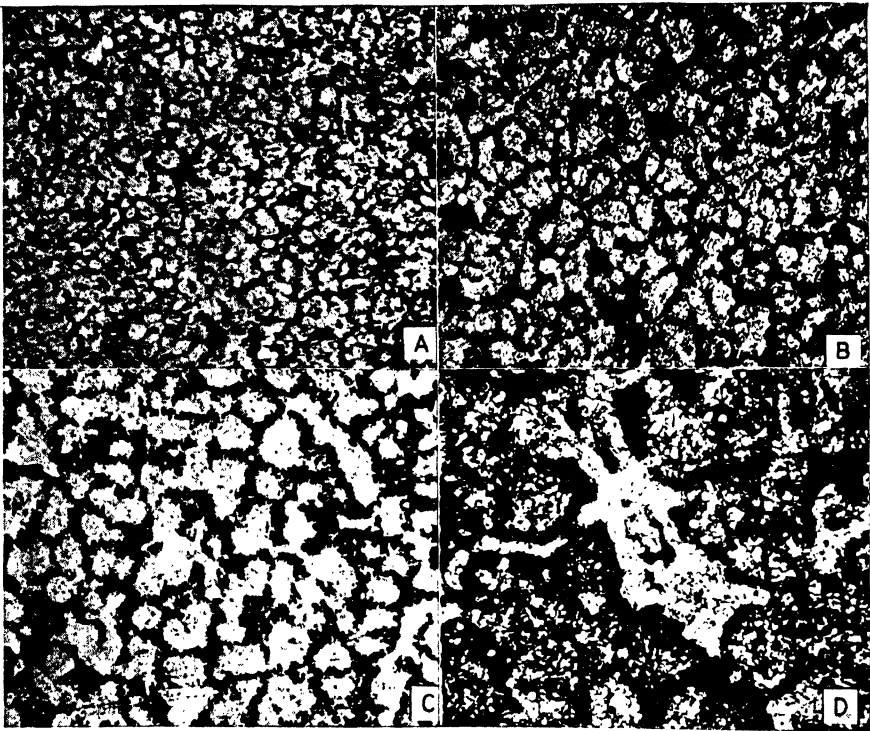
FIG. 277.—Effect of mass on the ferrite structure and distribution of a 1650°F. anneal, 0.27 per cent carbon cast steel. (a) Center 1-in. section, × 25; (b) center 2-in. section, × 25; (c) center 4-in. section, × 25; (d) center 8-in. section, × 25.

One of the functions of annealing is to relieve residual stresses developed in the casting by hindered contraction caused by mold construc-

TABLE LXXVIII.—MECHANICAL-PROPERTY VALUES NORMALLY EXPECTED IN THE MANUFACTURE OF CAST CARBON STEELS IN THE AS-CAST AND FULL-ANNEALED CONDITION[20]

Carbon content, per cent	Tensile strength, p.s.i.		Yield point, p.s.i.		Elongation in 2 in., per cent		Reduction of area, per cent		Charpy impact, ft.-lb.	
	As cast	Full annealed	As cast	Full annealed	As cast	Full annealed	As cast	Full annealed	As cast	Full annealed
0.20	67,500	67,000	37,000	38,000	30	31	48	52	20	23
0.30	80,000	78,500	40,000	46,000	19	26	31	41	14	20
0.40	91,000	89,000	41,500	51,000	15	21	24	32	8	15
0.50	98,500	96,000	42,500	52,000	11	17	18	24	3	7

tion or casting design. After these stresses have been removed, the casting is cooled in such a manner that new stresses are not added.

Normalizing Heat-treatment and Structures.—The normalizing heat-treatment is employed for the same reasons as for the annealing heat-treatment, except that by the faster rate of air cooling from the heating temperatures it is possible to reduce grain size and increase certain property values over those obtained by annealing. In general, normalizing produces higher yield and ultimate strength than full annealing. Ductility is approximately the same after either treatment, but normalizing often gives the steel higher resistance to impact. Full annealing results in a softer steel and greater freedom from stresses.

The nature of the mechanical properties obtained by normalizing can best be seen by referring to Fig. 325. Complete information on heat-treatments of carbon and alloy cast steels with corresponding properties may be found in the "Steel Castings Handbook."[20] Normalizing, as contrasted with annealing, increases the strength properties of alloy steels without much loss in the ductility properties. Representative values of a low-alloy steel may be best observed by studying Table LXXVIII. The normalizing treatment is the one usually given to low-alloy cast steels, since this treatment brings out the effectiveness of the alloy additions. The benefits of alloys in steel are obtained through the effect of their response to heat-treatment, such as a slower critical cooling rate or resistance to softening during tempering. Full annealing fails to take advantage of these benefits and is seldom used for alloy steels except as a first treatment to be followed by normalizing or quenching. Some alloy steels attain the optimum combination of properties through normalizing. This is especially true of the fine-grained steels that have great resistance to grain growth.

For castings in which high hardness is not demanded, normalizing is preferred to liquid quenching because it is a simpler process, gives less trouble from warpage, and is used on castings of complex design. Air cooling after the homogenization treatment constitutes a normalizing treatment. The heating temperatures and times at heat for carbon and alloy steels are given under the section dealing with homogenization.

A normalizing treatment may follow a full anneal or a preliminary normalizing treatment. The heating temperature for this second treatment should not greatly exceed the Ac_3 temperature. A temperature range of 1525 to 1550°F. is normally selected as the heating temperature for carbon steel, while 25 to 50°F. above their Ac_3 range is used for alloy steels. The holding time at this temperature should be longer than required, to ensure uniform heating and complete solution of the carbides in the gamma iron. Some of the alloy carbides such as molybdenum carbide are less easily soluble than is iron carbide and require a longer time

for solution. In most cases $\frac{1}{2}$ to 1 hr. after reaching a uniform temperature is sufficient. Since diffusion has been accomplished in the prior heat-treatment, the purpose of the second treatment is to obtain grain refinement without permitting opportunity for grain growth. A second normalizing treatment may not be necessary in some cases, if grain-size-control additions are made to the steel to prevent grain growth in the first normalizing treatment. A typical structure as developed by normalizing and double normalizing is shown in Fig. 278. Other normalizing structures may be seen in Figs. 269 and 280.

(a)　　　　　　(b)　　　　　　(c)

(d)　　　　　　(e)　　　　　　(f)
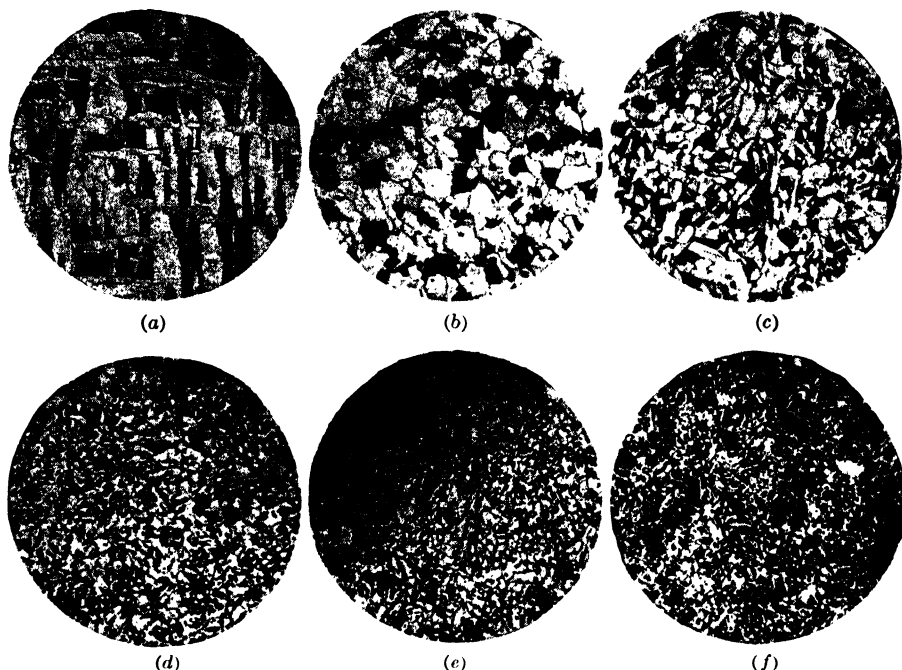
Fig. 278.—Comparison of the structures of (a) as-cast, (b) annealed, (c) normalized, (d) double normalized, (e) double normalized and tempered, and (f) quenched and tempered carbon cast steel.　× 75.

Since air quenching usually sets up cooling strains responsible for permanent stresses acting on the casting, it is desirable to temper the castings by reheating to a stress-relieving temperature. Ordinarily, no change occurs in the microstructure as may be seen in Fig. 278.

The ferrite-network type of structure is generally obtained for the normalizing grades of alloy steels. Structures similar to those shown in Fig. 269 are typical of this group of steels. It has been stated that after homogenization the original structure appears quite easily, particularly if the cooling is slow. For example, a slow cooling will produce large ferrite islands (Figs. 274 and 275), often arranged in a dendritic pattern such as in Fig. 274b. Cooling at a faster rate will produce ferrite-network

structures such as illustrated in Fig. 276. Upon air cooling, a fine ferritic-network structure is developed such as may be seen in Fig. 269. Thus the rate of cooling is instrumental in preventing the formation of the original structure. Moreover, by homogenizing at higher temperatures and thus coarsening the austenitic grain, the original structure is not so easily revealed. Likewise, the carbon content of the steel is instrumental in the ease with which an original structure may be developed. Cast steels of low carbon contents must be cooled rapidly, to avoid inheritance of the original structure.[7]
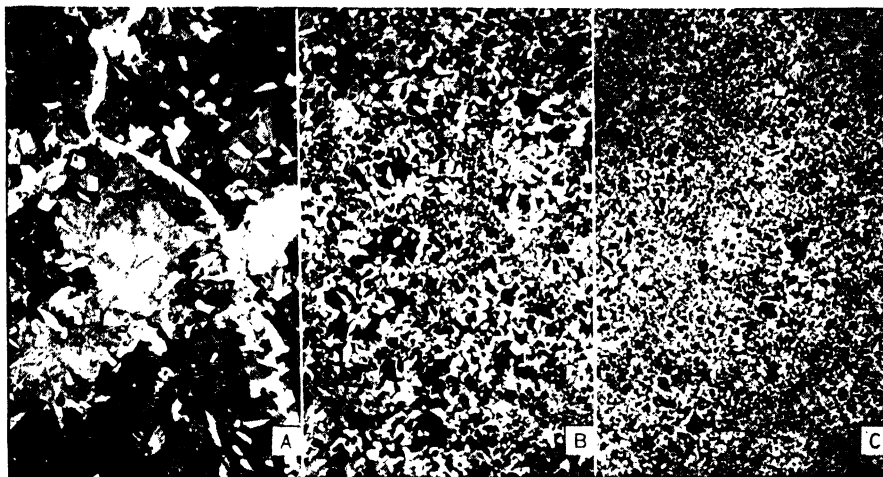
Fig. 279.—Microstructure of manganese-vanadium cast steel. Carbon 0.35 per cent, manganese 1.07 per cent, vanadium 0.12 per cent. × 70. (a) As-cast; (b) normalize 1650°F; (c) normalize 1650°F., normalize 1550°F.

Very fine-grained structures may be obtained on normalizing and double normalizing in certain of the alloy steels. A typical example of a fine-grain structure is shown in Fig. 279.

Liquid Quench and Temper Heat-treatment and Structures.—Steel castings are given a liquid quench and tempering heat-treatment when exceptional mechanical properties are required. Such castings will have a high degree of structural refinement and will be more resistant to wear.

Accelerated cooling such as water quenching will result in the following for steel castings:

1. Improvement in the strength-ductility ratio
2. Improvement in the yield-tensile ratio
3. Increase in the efficiency of alloys
4. Decrease in the notch sensitivity (increase in the impact strength)
5. Improvement in the properties in heavy sections
6. Increase in the hardness

Carbon steel castings respond nicely to the quenching and tempering treatment, particularly when the carbon content of the castings is above 0.25 per cent or when manganese is high (0.85 to 0.95 per cent). Excellent strength properties can be obtained and toughness can be increased in light sections. Carbon steels are, however, shallow-hardening and even in the medium carbon ranges (0.25 to 0.40 per cent) only a relatively thin shell is effectively hardened.

Most alloy steels require quenching and tempering to develop their best properties. Alloy steels are also quenched when high hardness is desired.

In many cases, the design of the casting is such that the drastic heat-treatment of liquid quenching cannot be used. Some castings that have intricate design or great differences of section are not quenched because of danger of cracking or warping. In those cases where it can be used, considerably improved properties may be obtained, as is shown by the example cited in Table LXXIX. Complete information on the quenched and tempered properties of carbon and alloy steels can be found in the "Steel Castings Handbook."[20]

TABLE LXXIX.—QUENCHING OF STEEL CASTINGS

Type	Carbon range, per cent	Quenching temperature, °F.	Time at heat	Quenching medium and temperature, °F.	Tempering	
					Temperature, °F.	Time at heat, hr.
Carbon	0.15–0.30	1600–1650	¼ hr. per in. of max. cross section	Water at 70	800–1250	2
Carbon	0.30–0.40	1550–1600		Water at 125	800–1250	2
Carbon	0.40–0.60	1500–1550		Water or oil at 125	800–1250	2
Alloy	0.15–0.45	25–50 above Ac₃ range	½ hr. per in. of max. cross section	Water at 125–150	800–1250	2

The quenching of carbon and alloy cast steels is preferably preceded by a normalizing or full-annealing treatment to avoid the coarse grain resulting from the long heating at temperatures necessary for homogenization.

After proper annealing, the castings should be heated uniformly to the quenching temperature. Conditions for the quenching of steel castings, including the time of heating at temperature and the quenching medium, are shown in Table LXXIX.

Straight carbon steels from 0.25 to 1.00 per cent carbon can be water-

or oil-quenched, depending upon the size and the tolerance allowed for distortion. Small-diameter pieces of high carbon are satisfactorily hardened in oil with no appreciable distortion. Alloy steels of a 0.35 to 0.40 per cent carbon are very critical and sensitive in regard to size, and much concern and thought must be given to the correct quenching mediums. To avoid cracking, care must be taken, based on the size and shape; light sections may be oil-quenched to procure the desired hardness to the axis. Medium sections from 1 to 2 in. cannot be handled satisfactorily with an oil quench. Sections over 2 in. can be safely quenched in water because of the larger mass of metal lowering the cooling rate, thus rendering the water quench less drastic. On all steels over 0.40 per cent carbon, an oil quench is recommended up to approximately 3½- to 4-in. sections, which will produce the depth-hardening properties. A complete discussion on quenching mediums has been prepared by Scott,[21] to which the reader is referred.

If the proper equipment and competent personnel are available, it is believed that the best method of quenching any and all castings, regardless of section thickness or complexity, is to use the time-quench procedure with water or brine as the quenching medium. Castings are removed from the quench while at temperatures of from 500 to 700°F., depending on the casting design and composition. This temperature can be correlated with the length of time the casting remains in the quenching medium, and the entire operation can be controlled by a stop watch.

Times of immersion in the quenching bath will depend upon the casting design and range from 30 to 90 sec. for small castings to upward of 15 min. for large castings. The castings are then removed and immediately placed in the tempering furnace. Castings are seldom allowed to cool in the quenching liquid to temperatures below 500°F. Cooling castings to temperatures much below 500°F. often results in excessive warpage or cracked castings.

Commercial steel castings (other than high-alloy castings) are not used in the liquid-quenched condition. Tempering temperatures range from 500 to 1300°F., depending on the properties desired. The heating time at the tempering temperature should be about 2 hr.

Figures 278 and 280 illustrate the structures obtained on quenched and tempered cast steels. These structures are compared with those obtained by other heat-treatments.

Cast steel compares favorably with wrought steel of similar composition as regards hardenability, and the same factors influencing hardenability of wrought steel influence cast steel in a like degree. It is suggested, therefore, that the reader turn to the classic works of Mehl, Grossmann, and Bain[22] for a complete analysis of this subject.

In quenched cast steel, hardenability is a function of the austenitic

grain size. The larger grains give greater hardness and deeper penetration of the hardening effect. Cast steel is inherently of coarse-grained austenite and therefore exhibits good hardenability. Even though
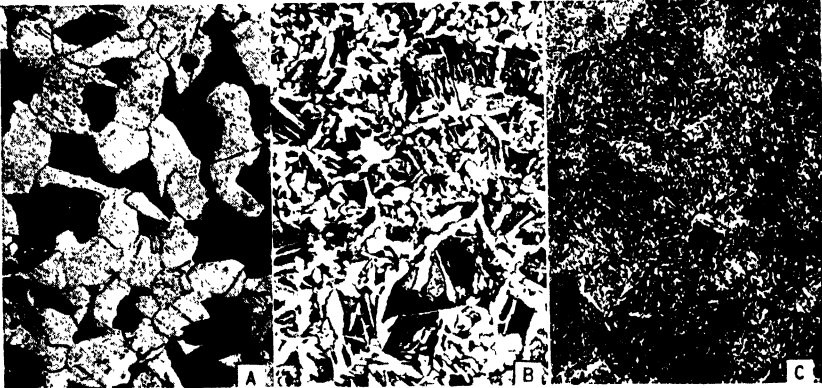
FIG. 280.—Comparison of structures in a 0.34 per cent carbon cast steel. (a) Full annealed, X 175; (b) normalized, X 175; (c) quenched and tempered, X 175.

aluminum is used in the deoxidation of most cast steels, the austenite grain size is large because of slow cooling.

Alloy additions will increase the hardenability of cast steels since the hardenability of a steel is increased by an element dissolved in austen-
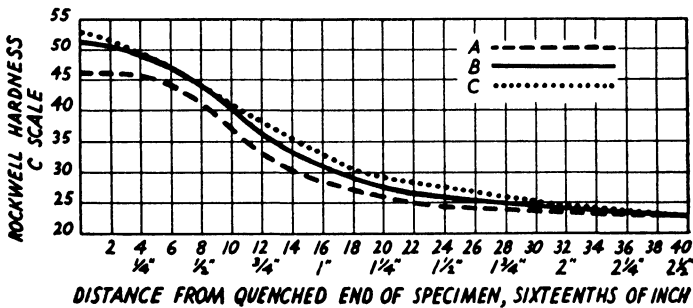
FIG. 281.—Hardenability curve for cast steel similar to 2330. (Eddy.[20])

Steel 2330	C	Mn	P	S	Si	Cr	Ni	Mo	Quench Temp., °F.	Grain size
A = cast............	0.28	0.69	0.043	0.028	0.41	0.12	3.30	0.03	1457	7-8
B = cast............	0.28	0.69	0.043	0.028	0.41	0.12	3.30	0.03	1560	7-8
C = wrought........	0.29	0.61	.....	.....	.....	.....	3.41	....	1505	7

ite, with the exception of a few such as cobalt and aluminum. Carbon, nickel, manganese, vanadium, molybdenum, and chromium are increasingly effective in the order mentioned.

The alloy additions to a carbon cast steel increase the time required for pearlite transformation and thereby increase the hardenability by making the action of the quench more effective.
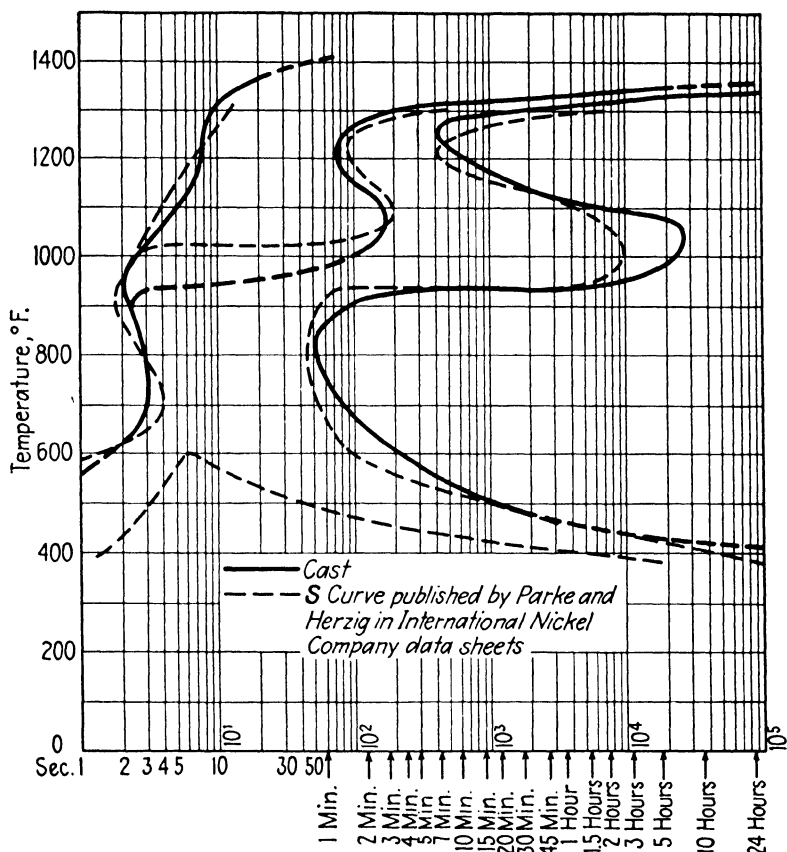
Fig. 282.—Comparison of the time-temperature transformation curves of cast and wrought steels similar to 4130. (*Eddy.*[24])

Type	C	Mn	Si	Ni	Cr	Mo	S	P
Cast....................	0.32	0.54	0.36	0.07	0.75	0.27	0.021	0.034
Wrought................	0.28	0.65	0.19	0.16	0.66	0.22	0.023	0.01€

Studies on the hardenability of cast steels[23,24] have shown that cast steel and wrought steel of similar composition have similar hardenabilities, as illustrated by the end-quench test (Fig. 281). Commercial cast steel will, in general, show slightly higher values than commercial wrought steel of the same carbon content, because of the higher silicon,

aluminum, and sometimes manganese contents of cast steel, since all cast steels are dead-killed steels. The hardenability of cast steel is dependent upon composition and grain size in exactly the same manner as wrought steel. An understanding of the properties developed in castings of varying sections can be obtained by the use of the end-quench test as a routine control in steel-foundry operations.

Since the end-quench hardenability curves as obtained on cast steels compare so favorably with those of wrought steels, it would necessarily follow that the isothermal transformations of the two types of steel would be similar. And such is the case since, on the basis of the results of the investigation of four selected cast steels, it was concluded by Eddy[24] that S curves for wrought steels can be utilized in the heat-treatment of steel castings. In Fig. 282 is illustrated a comparison of cast and wrought steels similar to S.A.E. 4130. The significance of the S curves and the effect of grain size, alloys, and other factors on the time required for the pearlite transformations may be obtained by reviewing the work of Grossmann and Bain[22] and Eddy.[24]

Tempering or stress-relief treatment is a heating operation applied to the steel casting after a normalizing or a liquid-quenching treatment, in order to relieve the strains and increase the toughness. Incidentally, it usually causes the casting to lose some of its hardness, but this is not the purpose of tempering; it merely goes hand in hand with it. Most liquid-quenching heat-treatments for steel castings are designed to promote toughness rather than hardness and wear resistance. Practically all foundries that do liquid quenching report that the heat-treating is carried out to definite hardness requirements. Satisfactory ranges (Brinell) have been tabulated[25] at various hardnesses. These are as follows:

Required	Range	Spread
475	444–514	70
300	269–321	52
200	187–229	38

It is well known that enormous internal strains may be built up in a casting during quenching. At times these strains are greater than the strength of the cast steel, and it cracks in the quench. Again, they may approach 98 per cent of the strength of the steel, in which case the casting does not crack during the quench but may break later upon the slightest provocation, as when the room temperature drops several degrees; such a drop may take place between day and night. The average strains are doubtless much below this figure, but the foundryman should remember that the useful strength of a casting is equal to the total strength of the

steel minus the internal strains. As an illustration, assume that a quenched casting has a total strength of 140,000 p.s.i. Suppose the unfavorable internal strains are 90 per cent of the figure, or 126,000 p.s.i. Then the remaining strength to do useful work is only 14,000 p.s.i. If by proper tempering or stress relieving these internal strains can be reduced to 70,000 p.s.i., the useful strength will be raised to 70,000 p.s.i., a gain of 400 per cent.

Tempering accomplishes its purpose through a combination of temperature and time. It is not sufficient merely to heat steel castings to
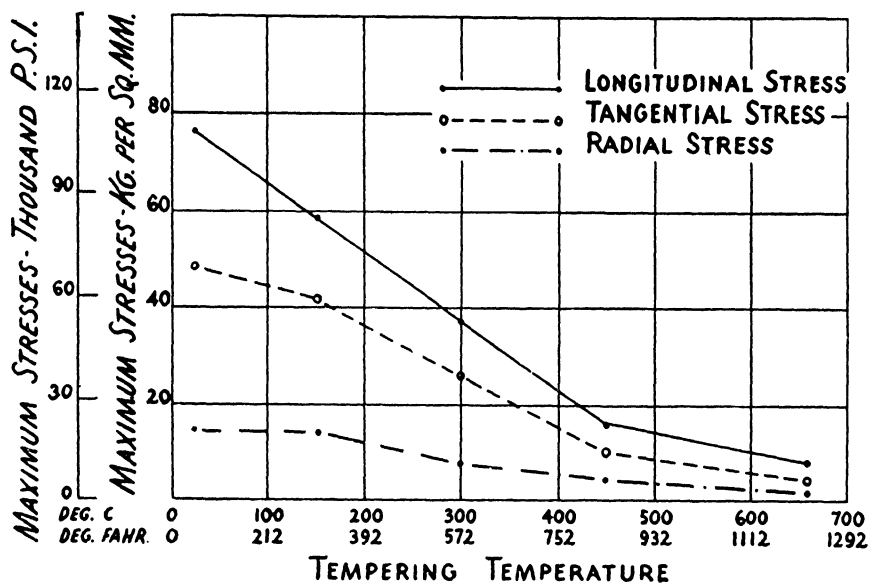
FIG. 283.—Effect of quenching and tempering temperature on the relief of internal stresses. (*Buhler, Buchholtz, and Schulz.*[26])

some definite temperature; they must be held at that temperature for a definite period of time.

The manner in which tempering relieves stresses induced in steel by quenching is illustrated[26] in Fig. 283. An 0.30 per cent carbon steel was quenched in water from 1560°F. and tempered at an ascending series of temperatures. It will be observed that a temperature of 1100 to 1300°F. is required in order to render the material virtually stress-free. Further conditions of the relief of stresses are shown in Fig. 276 and discussed under Annealing Heat-treatment.

The time that the castings remain at the tempering temperature depends upon the amount of stresses to be relieved, since time at temperature is instrumental in the relief of stresses as shown in Fig. 272. The relief of stresses ordinarily requires a longer time than the development of the proper tempering structure.

Most castings are furnace-cooled from the tempering temperature to 500 to 750°F. and then removed and air-cooled.   However, it has been definitely established that certain types of low-carbon-alloy steel give better impact values if quenched from the tempering temperatures than if cooled in the furnace.

Control of Grain Size.—The formation of austenite[27] begins in those regions of the dendritic pattern that have the lowest Ac_1 transformation. These regions lie between the branches of the dendrites or between the dendrites themselves, depending upon which area contains the least
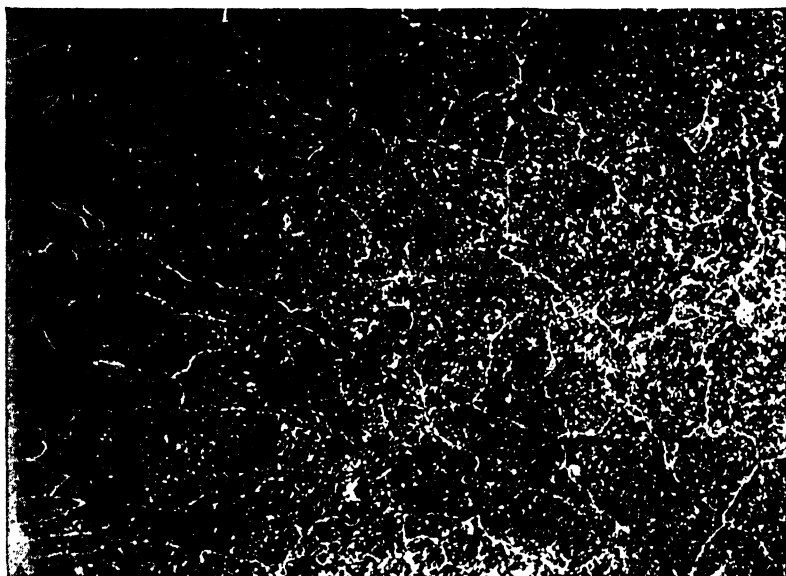
Fig. 284.—Austenite grain size.   Section of a 6-in. cube of medium manganese steel in as-cast condition.   × 1¼.   (*Courtesy of C. E. Sims.*)

number of sulphide inclusions.   Larger austenite grains are formed in these areas than in the other parts of the dendritic pattern.   It is these larger grains that coarsen first upon heating.

The austenite grain size in cast steel in the as-cast condition is large, as may be seen in Fig. 284.   The ferrite that forms from the primary austenite appears as large plates, as in a Widmanstätten structure or in a coarse network.   When the cast steel is reheated in the homogenization process and austenite is again formed, it is not surprising to find that the size of the austenite grains bears a relationship to that of the prior ferrite grains.   In the medium and lower carbon cast steels, it has been observed that the finer the ferrite before heating, the finer will be the subsequent austenite-grain size.

Different conditions of homogenization (temperature and time) pro-

duce different sizes of austenite grain.   As the homogenization temperature is increased, the austenite grain coarsens.   The coarsening does not begin until a temperature higher than that at which austenite forms is reached.   This higher temperature at which grain growth sets in may vary from 1600 to 2100°F.   Each cast steel may have a different coarsening temperature, which depends in general on the steelmaking method.

The composition of the steel has an influence on its grain size, although the carbon content has but little effect.   Chromium, nickel, and copper influence grain size and assist in grain refinement.   Molybdenum can also greatly influence grain refinement.   Probably the most effective elements in grain-refining tendencies are vanadium and aluminum, with the former the more potent of the two.   Titanium also acts as a grain-refining ele-

a   *b*
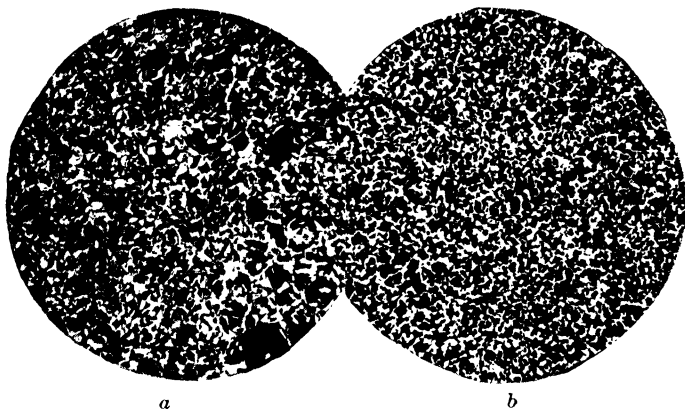
FIG. 285.—Effect of vanadium on grain size.   (*a*) A 0.38 per cent carbon cast steel, normalized and tempered; (*b*) a vanadium cast steel, carbon 0.38 per cent, vanadium 0.17 per cent, normalized and tempered.   × 75.

ment and even though it is not so effective as vanadium or aluminum, it is probably more effective than molybdenum.   The addition of small quantities of aluminum, so that the steel will contain 0.01 to 0.05 per cent residual metallic aluminum, changes the steel from a coarse-grained to a fine-grained steel, with entirely different heat-treating characteristics and mechanical properties.

The effect that vanadium has in producing a fine grain size can be seen in Fig. 285.   The photomicrograph marked *A* is a 0.38 per cent carbon steel, while *B* is of similar carbon content but contains 0.17 per cent vanadium.   Both cast steels have been given a normalizing and tempering heat-treatment.   Of the two photographs, *B* is decidedly finer grained.

The steelmaker is much interested in the methods of control of grain size and the mechanism of the inhibition of grain growth.   In general, most authorities are in agreement that fine-grained austenite is the result of inhibition of grain growth by properly dispersed insoluble

particles. Aluminum, vanadium, and titanium are the principal alloys used to produce fine grain size.

The influence of aluminum, vanadium, and titanium upon the structure of cast steel is first observed in the as-cast condition. Generally speaking, these alloys produce a less marked dendritic segregation and greater freedom from Widmanstätten patterns. This finer grain structure will persist throughout the various heat-treatments, the result being a finer grain austenite as well as a finer ferritic grain. Alloys are responsible for a sluggish allotropic change, thereby producing conditions of supercooling. Crystallization that takes place following supercooling occurs suddenly with the formation of many very small crystals as compared with fewer and larger crystals in the absence of supercooling.

It is apparently true that the grain refinement caused by an alloy will be very nearly proportional to the elevation of the allotropic transformation above the true critical temperature of the steel when it is heated.

A uniformly fine-grained cast steel gives excellent serviceability and has considerable shock-resistance ability. The impact value is greatly affected by the fineness of grain size and seems to be a more sensitive test of the heat-treatment of a casting than the tensile test.

An example is given by Sims[9] of the effect of grain size on the mechanical properties of a medium-manganese-vanadium cast steel, as follows:

	Coarse-grained	Fine-grained
Tensile strength, p.s.i...............	98,240	101,400
Yield point, p.s.i...................	63,060	66,210
Elongation, per cent................	27.0	27.5
Reduction of area, per cent..........	55.7	57.5
Izod impact, ft.-lb..................	18.5	48.0

The only difference between these two steels is that one was given the correct heat-treatment while the other was given an incorrect one. It will be noted that while the strength and ductility are nearly identical, the coarse-grained steel shows an Izod value of only 40 per cent of the fine-grained steel.

A fine-grained steel is not always desired. A large grain size gives greater hardenability and machinability. Also, where high-temperature strength is needed, the large-grained structure is superior.

HEAT-TREATMENT MODIFICATIONS

Besides the usual heat-treatments of annealing, normalizing, quenching, and tempering, there are a number of treatments that are based on these universally used heat-treatments but are modifications of the

methods and are used to varying degrees in the heat-treatment of steel castings. These modifications are explained in the following paragraphs.

Spheroidizing.—The term "spheroidizing" refers to the heat-treatment that produces a spheroidized form of the carbides present in the steel. It is obtained by heating lamellar pearlite to just below the critical temperature where it tends to spheroidize. This heat-treatment is often known as a "high-temperature draw" or "subcritical annealing." A spheroidized structure can be obtained by heating an annealed, normalized, or quenched casting to a temperature just below the critical, or it may be obtained in some types of steel by holding the casting at temperatures just below the critical during the cooling of the casting from the homogenization temperature. A spheroidized structure is developed primarily for greater machinability of cast steel.

The time at temperature should be equivalent to approximately 1 hr. per in. of section. The rate of cooling after a spheroidizing treatment has no influence on the hardness or the structure, providing it is not a water-quench treatment. Such a treatment does influence the steel somewhat, as may be observed under Precipitation Hardening.

Step Normalizing.—Step normalizing, followed by tempering, is used occasionally as a treatment for certain alloy steels. Particularly those containing chromium, molybdenum, and tungsten are step-normalized. It is a treatment in which the steel casting is held for a considerable period just below the critical temperature, quickly moved into the critical range for a short time and air-cooled, and then tempered. It is claimed by Malcolm[28] that by this treatment network structures are broken up with maximum refinement of grain, retaining high strength combined with good ductility and toughness. The treatment is given after homogenization treatment. A very fine ferritic-grain size is claimed for the step-normalizing treatment.

Quenching Following Differential Heating.—Castings may be differentially hardened by a method that involves quenching of the casting following differential heating. The process embraces the use of the oxyacetylene flame or induction heating to raise specific sections of the steel casting above the Ac_3 transformation point, so that subsequent quenching will produce a desired hardness and structure.

The flame-hardening process is limited primarily to surface heating of the casting, and a hardened surface layer can be obtained that may vary in depth from a mere skin to as much as $\frac{1}{4}$ in., according to the practice used and the composition of the steel treated.

By the induction heating method entire sections of castings as well as the surface can be hardened. In this process the heated portions may be hardened while maintaining the original properties in the unheated portions. Although this process has many advantages over flame

hardening, it requires elaborate and costly equipment and therefore is not applicable to the product of the jobbing steel foundry except in cases where mass production of certain design is assured.

Flame hardening may be a hand or a machine operation but is usually the latter. Multiple flame heads generally are provided with water jets, and the flame tips are set close together. The steel is water-quenched immediately behind the flame tip, as shown in Fig. 286. This process is a favorite one for hardening teeth on cast-steel gears of all types. Teeth are hardened on both wearing faces, as shown in Fig. 287.
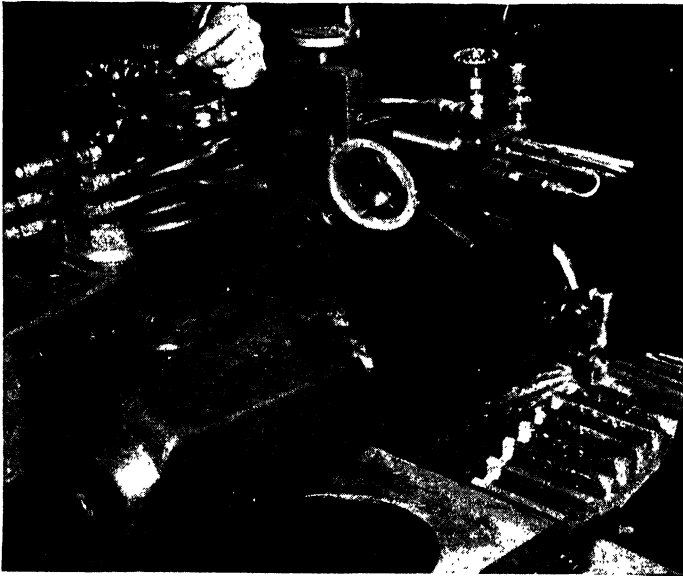
FIG. 286.—Flame hardening of a cast-steel gear. (*Courtesy of Linde Air Products Co.*)

The depth of the case varies from $\frac{3}{16}$ in. on the tip of the tooth to $\frac{1}{4}$ in. at the pitch line. Hardnesses of from 600 to 700 Brinell are produced. Castings subject to flame hardening usually have a previous normalizing and tempering treatment to produce the requisite toughness and ductility in the base sections of the casting.

Carefully controlled differential heating, followed by an accurate time quench, is consistently producing steel castings having as many as three Brinell hardness ranges in the same casting.

Differential Quenching.—In this process certain sections of a heated casting are subjected to a liquid quench. Differential quenching is obtained by spraying, spinning, or partial immersion in the quenching medium, and considerable ingenuity is exercised in the construction of fixtures to perform these various operations. Castings are ordinarily given a normalizing or annealing treatment and then heated to tempera-

tures normally employed for the liquid-quenching treatment. The casting is removed from the furnace and partially quenched, while the remainder of the casting air-cools.

Cast-steel wheels of various types in considerable numbers are given the differential-quenching treatment, the rim face being hardened by rotating the rim portion in water or by water-spraying the rim face. The quenching requires careful attention and is often a time-quench process.

Austempering.—A heat-treatment not yet applied to steel castings is austempering. The process consists of heating to a suitable temperature
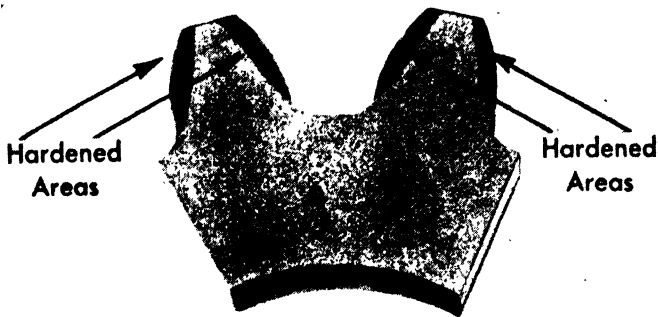
Unretouched view of etched cross section
of a flame-hardened gear tooth.

Fig. 287.—Unretouched view of etched cross section of a flame-hardened gear tooth.
(*Courtesy of Linde Air Products Co.*)

above the critical range and quenching in a liquid held at 300 to 800°F., depending upon the results desired. This results in the direct transformation of the austenite to the desired metallurgical constituents without the transformation to martensite followed by tempering, which is the normal quenching practice. This process, at a definite hardness, results in a very pronounced increase in the reduction of area and shock toughness. One of the disadvantages at the present time is that it requires a very careful selection of steels and can be applied only to the thin sections up to $\frac{3}{8}$ in.

Precipitation Hardening.—The strength and hardness of certain cast steels can be increased by a precipitation hardening or a quench-aging treatment.

Precipitation hardening is a simple process and is best known and used in the heat-treatment of copper cast steels. A little more than 1 per cent copper remains in solid solution in low-carbon steel at the completion of the gamma-alpha transformation (about 1290°F.). If subsequent cooling to room temperature is slow, most of this copper will be

precipitated from the solid solution, only about 0.35 per cent remaining in solution at normal temperatures.   If, however, the rate of cooling is increased, an increasing proportion of copper will be retained in solution until, at a rate of about 50°F. per min. or greater, no precipitation will occur, the steel becoming supersaturated with copper.   This super-saturated condition is stable at normal temperatures; but if the steel is reheated to about 750°F. or above, copper begins to precipitate out of solution in the form of submicroscopic particles.

The degree of hardening obtained from this precipitation of copper will depend upon the temperature at which precipitation occurs, the time at heat, and also the composition and previous heat-treatment of the steel.

The general practice for the precipitation hardening of copper cast steels is to employ a steel of from 0.15 to 0.30 per cent carbon content and from 1.25 to 1.50 per cent copper.   The castings are air-cooled following homogenization and then are reheated to the precipitation-hardening temperature of about 900°F. for about 4 hr., a minimum heating time.   The maximum increase in strength is produced at about 750°F., but the time required to produce maximum precipitation hardening is as much as 50 hr.   A shorter heating time with a correspondingly lower increase in strength is considered the most satisfactory from the standpoint of production.   Where impact strength is a major factor in specification requirements, the usual practice recommended is to age-harden at 1000°F.

TABLE LXXX.—EFFECT OF PRECIPITATION-HARDENING HEAT-TREATMENT ON A COPPER-MANGANESE-SILICON CAST STEEL

Finlayson[29]

	Annealed	Normal-ized	Normalized and precipitation-hardened
Tensile strength, p.s.i	90,750	114,000	129,500
Yield strength, p.s.i	71,000	95,000	115,000
Elongation in 2 in., per cent	25.5	19.0	17.0
Reduction of area, per cent	41.5	29.7	29.4
Brinell hardness number	187	257	292

C, 0.20; Cu, 1.86; Mn, 1.15; Si, 1.00.

The manner in which precipitation hardening affects the mechanical properties is shown in Tables LXXX and LXXXI for two copper alloy cast steels.

The advantages of precipitation-hardening heat-treatment as a method of improving the mechanical properties of cast steel are summarized as follows:[31]

1. The hardening is carried out at low temperature and over an extended period of time and therefore is not associated with abrupt changes of volume or structure such as occur in a quenching treatment. It is thus not accompanied by distortion or risk of cracking because of the setting up of internal stress. In fact, the treatment relieves residual casting stresses.

TABLE LXXXI.—EFFECT OF PRECIPITATION-HARDENING HEAT-TREATMENT ON A CAST COPPER STEEL

Greenridge and Lorig[30]

	Normalized	Normalized and precipitation-hardened
Tensile strength, p.s.i............	98,000	115,000
Yield strength, p.s.i.............	66,000	84,000
Elongation in 2 in., per cent......	22	19
Reduction of area, per cent.......	45	45
Brinell hardness number........ .	170	200

C, 0.30; Cu, 1.25; Mn, 0.75; Si, 0.40.

2. Precipitation hardening may be effected, if desired, during the later stages of machining owing to the absence of distortion at the comparatively low temperatures at which it is carried out.

3. Increase in hardness resulting from precipitation of copper is not accompanied by any appreciable reduction in machinability, such as would follow from an increase in hardness produced by raising the carbon content.

"Quench aging" is the term used to denote the changes taking place in steel after a final operation consisting of rapid cooling from an elevated temperature. This behavior is best observed in low-carbon steels after quenching. It has been shown[32] that carbon cast steels will precipitation-harden by quenching in water from below the critical range and aging at a low temperature.

A 0.25 per cent carbon steel will increase in tensile properties over those recorded in the annealed state upon quenching from just below the critical range and aging. Values are given as follows:

	Tensile strength, p.s.i.	Yield point, p.s.i.	Elongation in 2 in., per cent	Reduction of area, per cent
Annealed at 1650°F................	79,000	55,000	25.5	40.2
Annealed, quenched from 1256°F. and aged 20 hr. at 122°F........	85,500	63,000	19.7	33.1

The manner in which the hardness may increase on accelerated aging is shown in Fig. 288.   The higher the aging temperature, the shorter the heating time, to obtain the maximum hardness.

The quench-aging treatment may be used as a low internal-stress heat treatment, which can be readily applied to very irregular and intricate castings where the expansion and contraction obtained by quenching castings through the critical range may result in cracked or badly warped castings.

COMMERCIAL HEAT-TREATING AND PRODUCTION SCHEDULING

The heat-treating of steel castings as carried on in most foundries is operated on a batch-lot basis in car-type or stationary batch-type furnaces.   Castings placed in these furnaces go through a cycle of opera-
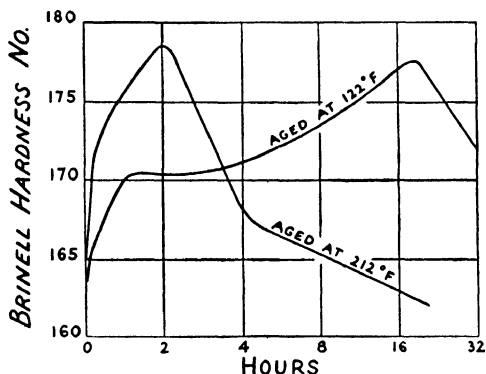
FIG. 288.—A 0.25 per cent carbon cast steel quenched from 1256°F. and accelerated aged at low temperatures.   (*Briggs.*[32])

tions that can be divided into three periods: heating, holding, and cooling periods.   These three periods will be considered individually.

Heating Period.—The heating procedure, according to most national steel casting specifications, shall consist of heating the castings to the proper temperature above the critical range.

Nearly all the information published on the subject of the heat-treating of castings is devoted to a discussion of the temperature requirements for heat-treating.   There is very little information available on the speed of heating of castings to reach this temperature.

Most specifications state that castings shall be uniformly heated at a controlled rate.   This would presumably mean that a rate that would not produce pronounced temperature gradients within the casting can be used.

The reason for specifying controlled heating rates is to avoid the setting up of harmful stresses due to temperature gradients in the castings that would cause the casting to warp or, in some cases, to crack.

The facts on how steel heats must be known obviously before an

estimation of temperature gradients and resulting stresses can be made. It is believed by many that the surface of a casting can be *at the furnace temperature* while the center of the casting is several hundred degrees cooler. The belief in cold centers is, of course, not true.

The outside surface of a casting continuously absorbs heat from the furnace gases until it reaches the furnace temperature. The center of the casting can obtain heat in only one direction—by conduction from the surface. As long as the center is colder than the surface, it will continue to draw heat from the surface, thereby cooling the surface and heating the interior. The only way for the casting surface ever to reach the furnace temperature is for the center to stop cooling it from below, and this happens only when the entire casting is uniformly heated. That is why the center and the surface of the casting arrive at the furnace temperature together.

For example, if a casting were placed in a furnace heated to 1600°F. and shortly thereafter the surface temperature of the casting were recorded and found to be 1000°F., it would be known that the center of that casting was at a temperature below 1000°F. However, when the surface temperature reached the furnace temperature of 1600°F., the center of the casting also would be at 1600°F.

The temperature difference between the center and surface of a steel casting depends on: rate of heating (or rate of heat transfer to surface); heat conductivity and specific heat of cast steel; thickness of the section; shape of the casting; and exposure of the surface (whether heated from one side, two opposite sides, or all sides).

It is possible to calculate time-temperature curves for the surface and for various points in the interior of a flat plate, cylinder, or sphere by strictly analytical methods. Grober[33] has done this and has plotted a series of curves that assist in the obtaining of temperature-gradient information. His curves show, for example, that a 36-in.-diameter ingot of low-carbon steel at 100°F. placed in a furnace at 2300°F. will have a surface temperature of 2235°F. when the center reaches a temperature of 2180°F., or a temperature gradient of 55°F. between the surface and its center.

Studies made by Trinks[34] on the heating of a 5¼-in.-square billet indicate that if a cold billet is placed in a furnace, the temperature of which is 2200°F., it will show a maximum gradient of about 250°F. when the surface temperature is at 1600°F. As the surface temperature approaches the furnace temperature, the gradient increases; when the furnace temperature is reached, the center of the billet is also virtually at furnace temperature.

If the furnace temperatures are not so great and the sections are small, the maximum gradients obtained between the surface and the

center are not so large. Palmer[35] showed a maximum gradient of about 60°F. at a surface temperature of 800°F. when a 3-in.-round steel bar was placed in a furnace at a temperature of 1450°F. He also showed that the gradient disappeared by the time the surface temperature of the bar had reached 1425°F.

The higher the furnace temperature, the shorter the time it takes the steel to reach that temperature and the greater are the temperature gradients that exist between the surface and the center of the steel object being heated. In the case of a furnace temperature at 1600°F., a difference of 100°F. may be obtained between the surface and the center during the early stages of heating, but the gradients are much less as the steel approaches the furnace temperature.

As the section of the casting increases in size, the time that it takes to reach a constant furnace temperature also increases. The following is an example of the effect of size on heating time when the furnace is held at 1450°F.

Size of steel bar, in.		Length of time to reach furnace temperature, min.
Circumference	Length	
¾	2	21
1½	3	42
3	6	85
6	12	170

The drastic rates of heating such as illustrated above are seldom, if ever, used. Castings are usually heated by increasing the furnace temperature and the temperature of the castings at the same time. However, recently some work was done[36] to establish the effects of high-speed direct heating with modern gas-air combustion on steel; it showed that a 2¼-in.-square billet could be heated to forging temperatures in 6 min. Even with such a brief heating time, only 200°F. was the maximum temperature difference between the skin and center of the billet. When the furnace charge is brought up to temperature with the furnace, there is very little temperature gradient between the surface of the casting and the center, providing, of course, that the furnace is being heated uniformly. Proper loading of well-cleaned castings, free of sand and excessive scale, is an important prerequisite of heat-treating. Castings should be so arranged that gases circulate freely around all sections and uniform heat distribution is assured. Typical thermal gradients in heating a 2- by 6- by 8-in. block of nickel-chromium-molybdenum cast steel are given in Fig. 289.

Another point should be considered in obtaining a uniform temperature heat-treating charge, especially with batch-furnace heat-treatment. In a charge of stacked castings, heat is mainly transferred along each casting rather than directly through the stack. In other words, the resistance to heat transfer at the surface of the casting is far greater than the heat resistance (*i.e.*, reciprocal of the conductance) of the steel itself. Thus a proper and uniform arrangement of castings to assume adequate circulation of hot gases is necessary for efficient batch heating.
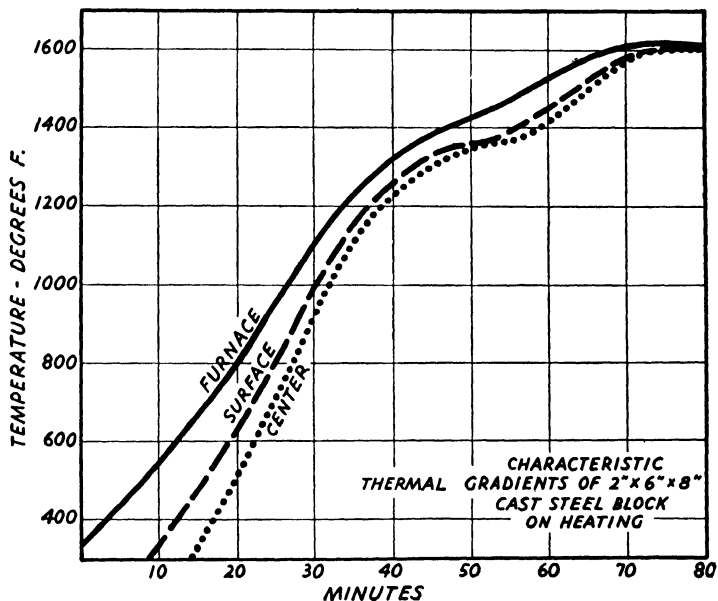
FIG. 289.—Characteristic thermal gradients for commercial heating of 2-by 6-by 8-in. cast-steel block. (*Juppenlatz.*[37])

Temperature gradients other than those found between the surface and center of a section may exist in the heating of a casting. For example, if a light, thin continuous section were attached to a heavy-section casting and the casting placed in a furnace heated at 1600°F., there is a distinct possibility that the light section would crack away from the heavy section. The reason for this is, of course, that the light section is completely heated through rather quickly and attains a high temperature, while the heavy section is at a relatively lower temperature. The difference in the rate and amount of expansion between the two sections sets up sufficiently high stresses to cause cracking. Castings do not need to be placed in a preheated furnace to have cracking or warping take place. Furnace-heating rates of a couple of hundred degrees per hour have been sufficient to cause cracking or distortion in alloy castings of complicated design.

It is believed that most cracking and distortion take place at the temperature and time that part of the casting is going through the critical-temperature range, while the heavier part of the casting is at a lower temperature.

The reason for this belief is that the tensile strength and yield point, which must be exceeded to cause cracking and distortion, are fairly high at the lower temperatures. While these values decrease as the steel temperature increases, nevertheless, the values below 1000°F. are sufficiently high to prevent cracking and distortion even in cases of very rapid heating. Also, temperature gradients existing in the casting at the critical-temperature range result in portions of the casting expanding while other portions are contracting with the setting up of higher stresses, which often result in distortion and cracking.

Thus the speed of heating can be accelerated at temperatures below the critical temperature, and rates of furnace heating of from 100 to 400°F. per hr. are not uncommon. In fact, for simple castings of nearly uniform section, rates of casting heating of from 1000 to 1500°F. per hr. can be carried on effectively without damage. In the latter case the castings are placed in furnaces that are at temperature or near the heat-treating temperature.

It is suggested that for most castings, with the possible exception of castings of a considerable varying cross section, the furnace temperature at the start of heating could be about 1000°F. or some high temperature below the critical-temperature range.

Upon cooling, castings may be removed from the furnace at 750°F., in accordance with certain national specifications. While the car is being unloaded and reloaded or while another loaded car is being shifted into place, the furnace can be heated without a charge. The new charge will cool down the furnace somewhat, since the car will take up considerable heat.

The furnace temperature could well be maintained at a temperature below the critical temperature until the temperature of the casting is equalized with the furnace temperature. Further heating may then begin. This procedure will save considerable heating time and will eliminate the possibility of great temperature gradients while the castings are traversing the critical range; consequently, the stresses will be kept to low figures, and warpage or cracking will be prevented.

Holding Period.—Most specifications require that the castings shall be heated "at the proper temperature above the critical range for at least 1 hr. per in. of thickness of the thickest section."

Thus the actual heat-treating temperature selected is delegated to the manufacturer. The consumer, however, has taken unto himself the specifying of the holding time. Since the time of holding is specified,

there is little or nothing that the manufacturer can do toward saving production time during this period of the heat-treating cycle.

Time may be saved in some instances if the manufacturer realizes that on heating a furnace at a rate of 100 to 200°F. per hr. the castings will be up to temperature shortly after the furnace reaches the heat-treat temperature. This is, of course, presupposing that uniform temperatures can be maintained in the furnace during heating. In this case the holding time can be considered to begin almost at once.

Recrystallization is complete as soon as the critical temperature is passed, but the rate of diffusion is rather slow at this temperature. The rate of diffusion increases rapidly with increases in temperature, and, in order to save time, a holding temperature somewhat above the upper critical range is used. Since the rate and extent of grain growth also increase rapidly, with temperature, too high a temperature is not desired.

Therefore, a compromise temperature, which is seldom more than 200°F. above the upper critical range but which will give diffusion in a reasonable time without excessive grain growth, is used.

Since the degree of microsegregation is increased by the lower freezing of heavy sections, the time required for diffusion varies with the section of the casting, castings with very light sections being homogenized in less than 1 hr. For heavier sections a rough guide is used that is usually stated as 1 hr. per in. of heaviest section.

There is, as a rule, no need to go to heat-treating temperatures of 1750°F. or above. Such a selection would mean that the heat-treating time cycle must be increased in heating and cooling the extra hundred degrees. Since 1 hr. per in. of heaviest section is specified as the holding time in most specifications, there is nothing to be gained by going to a higher temperature when temperature-time-quality requirements can be met at temperatures of approximately 1600°F. for most compositions.

If specifications make no reference to a holding time equivalent to 1 hr. per in., then the homogenizing temperature could be increased and the holding time reduced. Certain steel types can be heated to higher temperatures than others without undue grain coarsening, and hence the high temperatures are quite satisfactory.

A study would, of course, be required to ascertain if the time saved by reducing the holding time would be greater than that employed in heating and cooling the furnace to the higher homogenizing temperature.

Cooling Period. *Annealing.*—Slow cooling, as required by the annealing heat-treatment, is defined and specified in most specifications, which, in general, state: Castings shall be cooled slowly in the furnace. No casting shall be removed from the furnace until the pyrometer indicates that the entire furnace charge has fallen to a specified low temperature.

These national specifications differ only in the maximum temperature

at which the castings may be withdrawn from the furnace. Specification requirements as to this low temperature vary from 750 to 300°F. The reason that there are a number of various temperatures specified is that there are no test data available on which to base the proper withdrawal temperature. The 300°F. as used by one specification-writing body is most certainly ultraconservative. Annealed castings are supposed to be in as nearly a stress-free condition as possible; that is why the slow cooling is specified. It has been the opinion of certain purchasers that if castings are withdrawn at temperatures in excess of 300°F., cooling stresses will be set up and warpage, especially after machining, may occur.

Metallurgists from the industry who have given this subject considerable thought believe that a temperature of 750°F. for casting withdrawal from the furnace is not too great. For example, a casting of varying section at a uniform temperature of 1000°F. was removed from the furnace and allowed to cool in air to room temperature. If the casting was free from stress at 1000°F., it would likewise be free from stress at room temperature, providing that during the cooling the casting was not plastically deformed.

Plastic deformation could take place, since the sections of the casting vary and may therefore cool at different rates. This would mean that one portion of the casting would be contracting at a different rate from another portion, or it might be said that one section is trying to pull away from another section. Since this is not possible, stresses are set up which, if large enough in magnitude, may exceed the yield point of the metal with the result that plastic deformation occurs. When the casting becomes uniformly cooled to room temperature, it is under stress because the section that is plastically deformed is too long or too short, and its attempt to change the position of the other sections is reflected in the occurrence of stressed members.

If the stresses arising during the cooling of the casting are less than the yield point of the steel at the particular temperature, then plastic deformation does not occur and the casting will be relatively stress-free.

Since temperature gradients within the casting are responsible for the stresses' arising and the magnitude of the stress, it is essential that wide gradients not be allowed to exist.

It is believed, however, that castings cooling in air from 750°F. cannot form temperature gradients sufficiently great to cause stress magnitude that would exceed the yield point of the steel.

It must be remembered that as the temperature of the casting drops, the yield point increases; hence a greater temperature gradient may be tolerated.

Preliminary test work tends to show that at the lower temperatures, temperature gradients must be wide before the stress formed can exceed

the yield point of the steel. Under practical conditions such temperature gradients do not exist.

It is suggested that foundries not making castings to a specification that requires a withdrawal temperature remove the castings from the furnace when it reaches 750°F.

The amount of time required for furnace cooling is considerable, and anything that can be done to reduce this time without affecting the quality of the product should be done. A 10-ton furnace completely closed will require in the neighborhood of 36 hr. to cool from 1600 to 300°F., whereas a 100-ton furnace will require nearly 100 hr. to cool to 300°F. These periods are unnecessarily long.

Most specifications state that annealed castings must be cooled slowly. Now, the term "cooled slowly" is a relative one, and it might be argued that any cooling that is carried on at a rate slower than that of normalizing may be considered as a slow cooling. Some specifications require that this slow cooling must be carried on in the furnace and knowingly permit the furnace doors to be open.

It would seem that a better method can be obtained for accelerating the cooling by the installation of blowers in the furnace bottom for the introduction of unheated air into the furnace. The air would cool the furnace car primarily and the atmosphere above the car secondarily. This is just what is desired, since under normal furnace cooling the car temperature is considerably behind that of the furnace atmosphere. These conditions of cooling are not equivalent to the faster air cooling. The blower could be turned on after the castings temperature fell below the critical temperature. The rate of temperature drop expressed in degrees per hour is considerably greater from 1200 to 300°F. than from 1600 to 1200°F. In fact, the furnace requires more time to cool from 700 to 300°F. than from 1600 to 700°F. If the cooling rate is accelerated from 1200°F. downward, considerable time may be saved in the annealing cycle. Care must be used not to cool so rapidly that wide temperature gradients are established in the various casting sections. Rates of 100°F. per hr. for small castings of not great section changes to 50°F per hr. for large furnaces and large castings should not be detrimental. It should be remembered, of course, that the air blast must not enter the furnace directly on the castings, and this is why it is suggested that the air enter beneath the car.

Stress Relief.—There are some specifications that permit normalizing only if followed by a stress-relief heat-treatment. Cooling from the stress-relief temperature is treated in the same manner as specified in the cooling of castings from the annealing temperature, and points discussed under that heading apply here. The saving in time and fuel, of air cooling vs. furnace cooling, is offset if it is necessary to temper after

normalizing in order to relieve cooling strains.   A tempering or stress-relief heat-treatment is required after the liquid quenching of all castings and may be required after the air cooling of steel castings.   It is suggested that unless the castings are of a complex nature, the tempering or stress-relief furnaces should be maintained at the tempering temperature. This usually can be done, for if the casting will withstand the quench, it will usually · withstand the following quick heating without further warpage.

Stress-relief heat-treatments seldom are detailed in specifications; however, in some specifications the process is completely specified as to heating and cooling rates.   In this case the only thing that can be
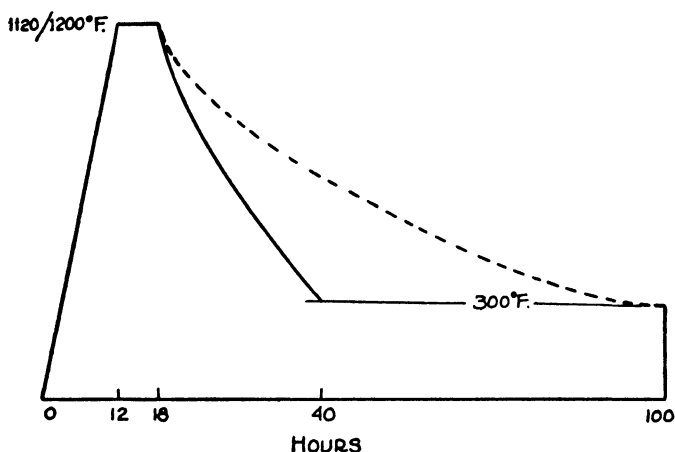
Fig. 290.—Typical stress-relief cycle showing saving in time accomplished through forced cooling.   (*Healey and Miller.*[38])

done is to take advantage of the fastest cooling rate allowed under the specification.

The manner in which a stress-relief cycle can be sped up is shown in Fig. 290.   A forced draft was installed in an 80- to 100-ton-capacity furnace, with the result that the cooling rate was increased from 11 to 45°F. per hr.   This in turn reduced the over-all time required to stress-relieve extremely heavy loads by approximately 60 per cent.   The shortening of the cycle, as shown in Fig. 290, makes it possible to increase the furnace turnover 150 per cent and provide reductions in heat-treating costs of 30 to 50 per cent.   The use of the shorter cycle expedites production and reduces the manufacturing time in cases where stress-relief treatment is necessary.

Normalizing.—Air quenching, or normalizing, as practiced by a number of foundries, violates a basic principle of heat-treatment, *viz.*, control of cooling rate.   Batch furnaces of the car type are often loaded with as many castings as the furnace will hold.   When the car is pulled

out into the air to allow the castings to cool, the sides and probably the top of the load are normalized; but the center of the load is probably annealed. This lack of control or air quenching can be minimized by loading the furnace very lightly, but this decreases the capacity of the furnace by as much as 75 per cent. If this is not done, then the castings should be removed from the furnace so that each casting can cool individually.

Even if castings are spaced properly for normalizing, the rate of heat extraction in still air is low and varies greatly with not only the section thickness but the shape and size of the casting. Structure and mechanical properties may vary widely for normalized castings, depending on the section thickness. Although the objection to normalizing is not overcome by water quenching, it is minimized, particularly in combination with the judicious use of alloys.

Quenching.—The fundamental metallurgical principles involved in the water quenching of steel casting[37] are no different from those applying to the quenching of any other steel structure when the complete hardening of the cross-sectional area is the aim of such treatment. The requisite factors for a successful quench are, in all cases,

1. Steel of sufficient hardenability for the section size involved.
2. A quenching bath of such capacity and rate of flow that the surface of the piece being quenched approaches the temperature of the quenching medium. A quenching bath of such nature provides the maximum cooling rate possible establishing the greatest temperature gradient between the center and surface of the quenched part.
3. Removal of the quenched part before its temperature has dropped below about 250°F. The risk of quench cracking is greatly increased if the temperature of the part is allowed to drop much below 250°F. before being tempered.

The factors that must be controlled and repeated time after time in as nearly the same manner as possible are

1. The temperature of the casting as quenched
2. The drasticity of the quench
   a. Water temperature
   b. Water circulation
3. The time of immersion in the water
4. The temperature and time of tempering after quenching

The manner in which the stated requisites for a successful quench are provided depends entirely upon the mass of the sections to be handled. The degree of freedom from quench cracking depends largely on the design of the part. It is most convenient first to discuss the effect of mass alone, without the complicating factors of design; this may best be done by the use of specific examples.

Rectangular blocks 8 in. wide by 10 in. long, with thicknesses of 1, 2, 3, and 4 in., respectively, may be used to illustrate the effect of mass on the factors involved in the water quenching of steel castings. It is at once apparent that the quench-cooling rates of each block will differ widely and that the analysis suitable for the 1-in. section may be entirely too low in alloying elements to enable the 4-in. section to quench out properly. It is also clear that a certain quench-bath capacity and rate of flow of quenching medium may allow a rapid quench for a 1-in. section and be quite insufficient for a heavier section.
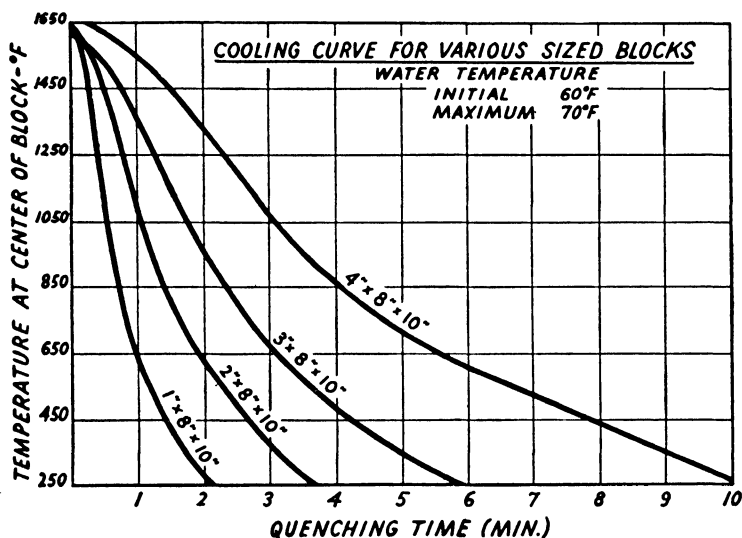
FIG. 291.—Cooling curve for various-sized cast-steel blocks. (*Phillips.*[37])

Figure 291 graphically illustrates the cooling rates at the center of sections of various thicknesses; it can be seen that it is impossible to quench light and heavy sections of the same analysis together without cracking the lighter sections or "slack-quenching" the heavier parts.

Figure 291 clearly shows an apparent but often overlooked fact. It is impossible to quench together, successfully, castings of widely varying section thicknesses. A practical solution of the problem is individual quench, with a specified length of time in the water for each section size. Reference to Fig. 291 shows that in order to avoid cracking the fully quenched 1-in. section must be removed from the water in about 2 min. after immersion, whereas in this same space of time the 4-in. section is still at a temperature of 1350°F. Where individual quenching is practiced, each casting may be handled through the quench in accordance with the full quenching of each piece without risk of quench cracking.

It is, of course, possible to batch-quench castings of different sections,

provided they are so handled that only castings of the same thickness are quenched together. Successful batch quenching requires good supervision and close attention to detail. It is imperative that the number and spacing of the castings in the load be a constant; furthermore, the relationship between quench-tank capacity, rate of water flow, and tonnage of castings quenched at one time must be a constant in order to obtain consistent quality. It is the amount of water that comes into contact with the castings, and not the total amount in the tank, that determines the severity of the quench. Practice has demonstrated that 3-ton batch loads may be fully quenched when the tank has a capacity of 2,000 gal. of water per ton of castings and the water circulates at a rate of 1,000 gal. per min. per ton.

The most serious difficulties encountered in the water quenching of steel castings arise as a result of objectionable design features. It is not mass itself that causes difficulty but rather the manner in which the designer joins sections of varying thicknesses.

It was pointed out that rectangular blocks of the same length and width but with thicknesses of 1 in. and 4 in., respectively, cannot be successfully quenched together; it is clearly demonstrated in Fig. 291 that under these conditions it is impossible to quench fully the 4-in. section without cracking the lighter piece. However, steel castings with sections varying from 5 in. to 1 in. are being successfully water-quenched in large tonnages; such castings have microstructures that are predominantly martensitic after quenching and are free from cracks. There are numerous other examples of castings involving extreme changes in section that are being satisfactorily handled through the quench; among such examples are cast parts weighing 85,000 lb. with sections changing from 6 in. to 3 in., and castings of a few pounds in weight with sections varying from 2 in. to $\frac{1}{2}$ in.

Several significant facts must be noted in connection with the examples given. The designs of the castings are such that the transition from one section to another is accomplished gradually; there are no sharp corners or sudden discontinuities of section. Further, the heavier sections occupy the largest area of the castings, and the lighter sections are so joined that heat flow from the more massive areas retards the rate of cooling to such a point that the thinnest section does not drop below 250°F. until the rest of the casting has reached a temperature in the neighborhood of 400°F. or lower.

Figure 292 illustrates the rate of cooling of a 2-in. section joined to a $\frac{1}{2}$-in. section. Comparison with Fig. 291 will show the effect of the $\frac{1}{2}$-in. section in accelerating the cooling rate of the heavier connected piece. The manner of joining the two sections is the poorest possible and is probably never encountered in practice. However, the heat conduction

from the 2-in. section through the ½-in. section during water quenching helps to maintain some heat in the lighter sections and also increases the cooling rate of the heavier section.

One method of equalizing the cooling of varying sections is that of blocking off various critical sections. If the design is such that there are thin sections that persist in cracking, even with timed or intermittent quenching, it is the practice to block these sections off in one of two ways. In some cases the position of the casting as loaded and quenched will allow the critical sections to be blocked off with fire clay, or fire clay and
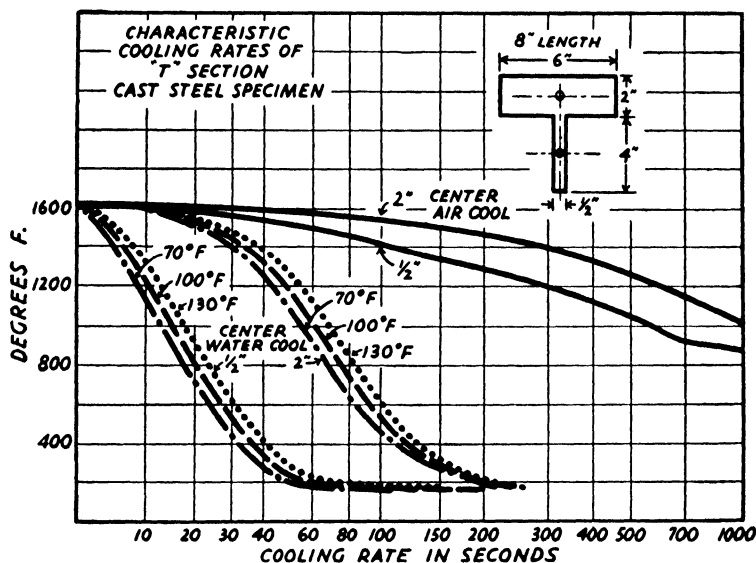
Fig. 292.—Characteristic cooling rates of T-section steel casting. (*Juppenlatz*.[37])

asbestos. In other cases it is necessary to make fixtures that fit around or on the critical areas. It has been found that in order to get effective results from blocking off, it is necessary to have perfect contact between the fixture and the casting section, and wherever possible these joints are sealed up with fire clay, or fire clay and asbestos.

Furnace Equipment.—All foundry operators should investigate the manner in which their furnaces heat to see that conditions of uniform temperature are obtained. Studies should be made of the gradients throughout the furnace during the heating and cooling period, as well as those that exist when the furnace is at temperature. If gradients of more than 50°F. exist, more efficient methods of firing should be provided or the furnace should be better insulated. The flame should not be allowed to be effectively felt in any area where the castings are located, since undue temperature gradients within the casting may result. The

installation of forced-draft cooling within furnaces appears to be a wise move, so that cooling rates can be controlled—thereby increasing production and usefulness of the furnace.

It is suggested that considerable thought be given to revising the long-established method of mass-batch heat-treating to make it a continuous method that provides more uniform heating. It is believed that the continuous furnace, besides enjoying the feature that castings can be more uniformly heated, offers an ideal arrangement for all heat-treatments following the homogenization treatment, since production times for the heat-treating cycle can be held at a minimum.

It is believed that the quenching and tempering of castings on an industry-wide basis will be realized in the near future. It is almost a certainty that specification requirements will be increased and that the quenching and tempering of castings can be most effectively handled from continuous heat-treating furnaces.

Most open-hearth foundries use pit- or car-type furnaces. Four open-hearth foundries have continuous automatically controlled furnaces. The hearth- or box-type furnaces are used almost universally, although they mainly supplement the car-type furnaces. The car-type furnaces, and those furnaces able to handle $1\frac{1}{2}$- to 150-ton loads, are the most popular in the industry. In four electric foundries continuous furnaces are used, all being the pusher type. There are only a few of the newer recirculating-air-type furnaces for tempering in use in foundries.

The types of fuel used for firing heat-treating furnaces in the steel foundries have been studied,[25] and the following percentages are reported:

	Per Cent
Oil	40
Gas	30
Oil and gas	15
Oil and electricity	6
Electricity	3
Oil and propane	2
Gas and electricity	2
Oil, gas, and electricity	2

There is a wide difference between foundries regarding the type and complexity of heat-treatments used, depending largely upon the type of casting produced. Some foundries liquid-quench and temper over 30 per cent of their entire production, whereas approximately 50 per cent do not quench any of their product. In many cases the latest heat-treating developments such as atmospheric control, recirculating-air draw furnaces, continuous furnaces, and flame hardening are employed.

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CHAPTER XV

THE WELDING OF STEEL CASTINGS

Cast steels and steel castings are welded by a number of different processes. These may be classified in the following manner for carbon and alloy cast steel:

Arc fusion.................	Most commonly used
Gas fusion.................	Commonly used
Thermite..................	Occasionally used but not in the foundry
Thermite pressure.........	Seldom used but not used in the foundry
Resistance................	Seldom used but not used in the foundry
Bronze weld...............	Occasionally used but not in the foundry

Since the only two processes being used in steel foundries in the United States are arc and gas welding, only these two will be considered.

Gas Welding.—In this process of welding the heat is supplied by the combustion of a suitable fuel gas with atmospheric or pure oxygen. The fuel gas and oxygen are mixed in a suitable torch and so regulated as to produce an impinging flame having the desired characteristics. The gases most commonly used are oxygen and acetylene, although hydrogen and other combustible gases are used to some extent. Ordinarily, a filler metal is added in the form of welding rod to make the welded joint.

In the combustion of acetylene and oxygen the resulting flame has characteristics that indicate whether it is reducing or oxidizing. If the acetylene and oxygen are in proportion, a neutral flame results. In most gas welding, the neutral flame is recommended; but in certain welding applications a slightly carburizing or slightly oxidizing flame is considered desirable. Certain of the alloy materials are welded to advantage with a slightly reducing or carburizing flame. An important characteristic of the oxyacetylene flame lies in the protection of the molten metal from atmospheric attack by the outer envelope of the flame. The surrounding oxygen at this point is consumed in the final combustion stage of the flame (Fig. 293).

Welding with the oxyacetylene torch is a procedure similar to the shielded-arc method. The source of heat is independent of the welding rod, and the sequence of events is controlled by the welder. The deposition of weld metal resolves itself on repeating the cycle of preheating the base metal to a semiplastic condition—and simultaneously preheating the welding rod to a temperature of not more than 1500 to 1800°F.

When the base metal reaches the desired temperature, the welding rod is advanced into the heat zone and a small volume is melted off and deposited upon the plastic base metal.

The drops melted off the tip of the welding rod are transferred onto the work partly by gravity but mostly by capillary attraction, since the welding rod is in direct contact with the deposited metal at the instant it is advanced under the inner cone of the oxyacetylene flame. As has been pointed out, the outer envelope of the flame provides a neutral atmosphere, which protects from atmospheric contamination the molten puddle of deposited metal and also the preheated end of the welding rod and the

Oxy-Acetylene Welding
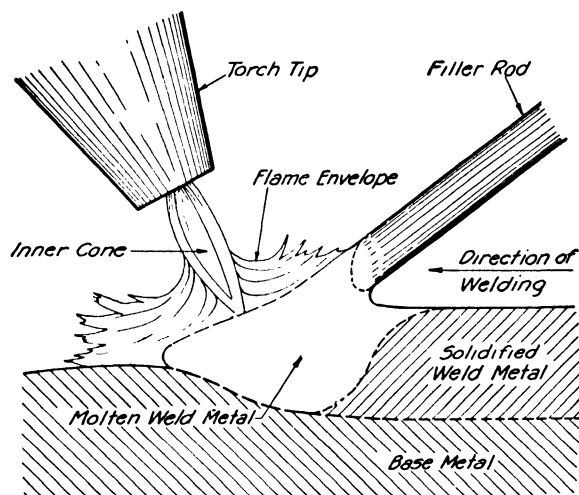
Fig. 293.—Oxyacetylene welding.

weld metal during the period it is melted off the rod and transferred to the work. The fact that the weld metal is deposited mostly by capillary attraction makes possible the use of gas welding in any position.

Arc Welding.—In arc welding the surfaces to be joined are fused by the heat of an electric arc. There are three forms of arc welding: metallic, carbon, and atomic hydrogen. These are listed in the order of preference and use in the steel foundry of today. Welding current for carbon arc welding is always from a d-c source. Metallic arc welding utilizes either direct or alternating current. Atomic-hydrogen arc welding generally is done with alternating current.

Metallic Arc Welding.—In this process the arc is struck between a metal electrode and the work. The heat of the arc fuses not only the work but also the electrode, which is in the form of a wire of suitable

material.   The fused-electrode material passes through the arc and unites with the fused base metal to form the welded joint.   There are two major classes of arc welding: unshielded and shielded.   The former term is applied when the electrode is either a bare wire or lightly coated with small amounts of flux.   In shielded metallic arc welding the metallic electrode is covered with a thick layer of fluxing ingredients with or without a fibrous binder.

Carbon Arc Welding.—In this process the arc is usually drawn between a carbon or graphite pencil or electrode and the work.   The heat of the arc causes fusion of the base metal, and, in case additional metal is required, it also fuses a filler rod, which is fed into the joint at the desired rate.   The process is not commonly used because of the erratic nature of the arc.   The arc can be controlled, however, by superimposed magnetic fields.   The filler rod may or may not be of the shielded type.

Atomic-hydrogen Arc Welding.—In this process an a-c arc between two tungsten electrodes in a stream of hydrogen gas furnishes the source of heat.   The arc column is defined by a thin, brilliant line that forms a loop under the influence of its own magnetic field and the streaming gas. In this arc column, which has a temperature between 6000 and 6500°F absolute, the hydrogen molecules are completely broken down into atoms, and during this process a great deal of energy is absorbed.   When this atomic-hydrogen flame, which is coincidental with the arc column, is placed in contact with the metal parts being welded, the dissociated hydrogen atoms recombine into molecules and give to the surface the heat that was absorbed during dissociation.   The weld metal may be supplied as surplus metal from the parts being welded or may be added from a filler rod.   The atomic-hydrogen arc is surrounded by an envelope of hydrogen gas, which protects the work and electrodes from oxidation.   The size of the atomic-hydrogen arc may be adjusted by the operator by changing electrode separation during operation; in this way an arc varying in size between $\frac{1}{8}$ in. and over 1 in. is obtainable.   The freedom from oxidation and the intense concentration of heat, combined with freedom from magnetic blowing effects, are features characterizing hydrogen arc welding.

In the following discussion the subject of welding will be treated primarily from the standpoint of the metallic-arc process, since it is by far the most favored process in use in the steel foundry.   Reference to the other processes will be made whenever special points require emphasis.

Weldability of Cast Steels.—The welding of cast steel presents the same problems as the welding of rolled steel of similar chemical composition.   Some individuals have claimed that, from the standpoint of welding, cast steel is more difficult to weld than rolled steel.   This is not a correct claim, and the statement probably means that the repair of defects usually involves welding in restricted locations, so that the weld

metal is able to contract in one direction only. Unless precautions are taken, contraction cracks may occur. However, such objections likewise apply to any welded fabrication.

There has been considerable discussion as to what constitutes a readily weldable steel. "Readily weldable" presumably means that no outstanding precautions must be taken before or after welding. In other words, the weldability of steel may be defined as the property that permits a localized fusing of metals by a welding process and that results in continuous bonding without any undesirable change in the mechanical properties of the material adjacent to the weld.

The American Society for Testing Materials has proposed tentative specifications for carbon steel castings suitable for fusion welding. These specifications allow the carbon content to be as high as 0.35 per cent with 0.60 per cent manganese, provided the casting is subsequently heat-treated after the welding operation. This grade of cast steel should also be preheated to 300°F. prior to welding. Castings that are not heat-treated following the welding operation are limited in carbon content to 0.25 per cent and in manganese content to 0.75 per cent. Thus it may be noted that the manganese and carbon contents are rather tied together. This is very apparent upon further study of the specification, wherein a reservation provides that for each reduction of 0.01 per cent below the maximum specified carbon content, an increase of 0.04 per cent manganese above the specified maximum is permitted up to a maximum of about 1.10 per cent manganese. The tentative specification for weldable carbon steel grades also calls for a maximum of 0.60 per cent silicon, 0.05 per cent phosphorus, and 0.06 per cent sulphur.

It is recognized that unspecified alloying elements may be present in cast steel, and in the interest of uniform welding the following restrictions (when specified) have been placed on the A.S.T.M. welding grades:

	Maximum Per Cent
Copper	0.50
Nickel	0.50
Chromium	0.25
Molybdenum plus tungsten	0.25
Total content of these unspecified elements	1.0

The Weldability Committee of the Engineering Foundation, commenting on the above A.S.T.M. welding requirements, pointed out that weldability by the electric-arc process is known to be a function of chemistry, material thickness, and temperature. Many types of test have been used to study weldability. Most of these tests have been described in the literature and some are known as

1. Bead-hardness test
2. T-bend test

3. Nick-bend bead test
4. "Lehigh Guide Book" method
5. Restrained-weld test

Each test gives certain specific data regarding weldability, but none can be recommended as a standard weldability test.

Although there are no standard weldability tests, the following may be used as comparative tests:

1. Bead-hardness test. Useful to determine maximum heat-affected zone hardness and tendency toward under-bead cracking.

2. T-bend test. Gives measure of the ductility of the heat-affected zone. Limited to fillet welds and plate thicknesses of about $\frac{3}{4}$ in.

3. Nick-bend bead test developed by the Naval Research Laboratory. Gives measure of ductility of heat-affected zone. Can also be used to furnish hardness data.

4. "Lehigh Guide Book" method. Based upon ductility of heat-affected zone.

5. Restrained-weld test. Measures tendency of base material to crack as the result of welding under restrained conditions.

Castings of 0.25 per cent carbon and 0.70 per cent manganese, without any adventitious elements, can be welded in any thickness without preheating, providing the initial temperature is not lower than 32°F. This grade of casting with maximum chemistry (carbon, 0.25 per cent; manganese, 0.70; silicon, 0.60; chromium, 0.25; molybdenum, 0.25; nickel, 0.25; copper, 0.25) might give trouble in heavy thicknesses without preheating. The heat-affected zone hardness would be in excess of 400 Vickers (384 Brinell) on sections 1 in. or more in thickness, and 350 Vickers (341 Brinell) is generally considered a safe top limit of hardness to ensure against under-bead cracking. Also, the calculated hardenability by Grossmann's method gives an ideal diameter greatly in excess of the tentative 1-in.-diameter limit proposed by Jackson, Pugacz, and Luther.[1]

In order to prevent welding difficulties on heavy sections, it is recommended that the adventitious elements of this grade of castings be limited as follows:

	Maximum Per Cent
Copper	0.50
Nickel	0.50
Chromium plus molybdenum	0.25
Tungsten	0.10
Total content of these unspecified elements	1.0

For each 0.1 per cent below the specified maximum alloy content of 1.0 per cent, an increase of 0.02 per cent chromium plus molybdenum content and 0.06 per cent nickel and copper content above the specified maximums will be permitted.

after welding. Even under the conditions of composite construction, wherein steel castings are welded to wrought steel, stress relieving is generally the accepted procedure. Thus, since conditions are more ideal, higher percentages of carbon, manganese, and other alloys may be used in the base metal, while what may be termed as a readily weldable steel is maintained.

Base Metal.—A number of experiments have been made on steel, both in the wrought and cast form, to determine its weldability. Broadly speaking, all tests for weldability are based on subjecting the material to the temperature cycle involved in welding and testing the ductility of the resulting structures. The thermal cycle is usually obtained by an actual welding operation, such as the laying of a bead weld. The ductility may be evaluated by means of some type of test such as bend, tensile, hardness, impact, etc.

A comparison of tests for the weldability of steel has been made by Jackson and Luther,[2] who conclude that the effects of a standard welding technique or a range of welding technique on a steel or a series of steels may be completely determined by an extended use of the following:

1. Chemical analyses
2. Test of mechanical properties (tensile and impact)
3. Test of actual weld joint such as the single V-groove weld
4. T-bend test
5. V-notched slow-bend test
6. Test of structure and bead hardness
7. Nick-bend bead test
8. Lehigh ductility test
9. Restrained-weld test

A broad view of the problem to be encountered in welding may be obtained from the chemical analyses and mechanical properties. The single V-groove weld, root, and face-bend specimens and restrained-weld test are practical ones. The T-bend test is excellent for the study of quality of welding steels. The V-notched slow-bend test, nick-bend, and Lehigh ductility tests are sufficiently sensitive and flexible to cover the entire range of variations encountered in welding procedures. Although a hardness survey by itself is of questionable value, it has a definite place as a control test for steels of known composition and of a standard quality.

If nine sets of tests are necessary to evaluate the weldability of a steel under ideal conditions, it can easily be seen that a knowledge of which steels may be considered readily weldable may be somewhat limited. In fact, Jackson and Luther made six sets of tests on 20 carbon steels and then failed to make any recommendations as to their weldability. The choice presumably rested on the conditions under which these steels were to be used. If it is considered that conditions are similar to those

that exist in the foundry, *i.e.*, minimum welding temperature 70°F., enclosed buildings, and down-hand welding, then the tests of Jackson and Luther show that a 0.30 per cent carbon steel with manganese content as found in steel castings is readily weldable.

Steels have been classified into three groups by Aborn,[3] based upon welding practice and the relative orders of maximum safe cooling rates (Table LXXXII).

TABLE LXXXII.—WELDABILITY OF STEELS
Aborn[3]

Steel types	Composition	General weldability	Pre-heating	Stress-relief annealing
I	Carbon steel, carbon under 0.30 Low-alloy steel, carbon under 0.15	Readily welded	None	None
II	Carbon steel, carbon 0.35–0.50 Low-alloy steel, carbon 0.15–0.30	Weldable with care	Preferable	Preferable
III	Carbon steel, carbon above, 0.50 Low-alloy steel, carbon above 0.30 or alloy above 3 per cent	Difficult to weld	Necessary	Necessary

In explaining this table, Aborn pointed out that it should be understood that the transition from one group to the next is gradual, and that marginal compositions possess in some degree the characteristics of the adjoining group.

Another classification of weldability has been made by Jackson and Rominski,[4] given in Table LXXXIII.

TABLE LXXXIII.—CLASSIFICATION OF WELDABILITY
Jackson and Rominski[4]

Grade	Average Brinell hardness	Preheat	Stress relief
I	Under 200	Unnecessary	Unnecessary
II	200–250	Slight, for thick sections	Thick (or thin) sections
III	250–325	300°F.	All sections
IV	Over 325	400°F.	Without cooling from preheat

It is generally agreed among welding engineers that to ensure absence of under-bead cracking in carbon steels of varying manganese contents, the Vickers Brinell hardness should not exceed 350 Brinell if the welded structure is to be stress-relieved after welding.

Studies on carbon-manganese steels, carried on at Battelle,[5] showed that steels with less than 0.25 per cent carbon and less than 0.50 per cent

manganese presented no problem from the standpoint of loss of ductility next to the weld.   Steels with more than 0.25 per cent carbon and slightly over 0.50 per cent manganese are borderline as to weldability when they are to be used in the as-welded condition, but they present no problem when stress-relieved.   This is true, regardless of section thickness. Steels having more than 1 per cent manganese and more than 0.30 per cent carbon may not be considered weldable without precautions such as preheating and stress relieving.   The Battelle work shows that the total of carbon plus manganese over 6 $\left(C + \dfrac{Mn}{6}\right)$ may be used as an index in the ranges of 0.20 to 0.30 per cent carbon and 0.50 to 1.0 per cent manganese.   The indicated carbon-equivalent limits for keeping below maximum Vickers hardness of 350 when welding at about 70°F. are as follows:

Section Thickness, In.	Carbon Equivalent for Hardness under 350
¼	0.65
½	0.55
1	0.45
2	0.40

The Naval Research Laboratory[1] work broadly confirms the tentative conclusions from the Battelle study, and further differentiates with respect to silicon content.   The presence of silicon is effective in reducing the manganese or carbon content permissible for straightforward welding.

The reason that base metal is so important in the welding process is that many welds that fail do not fail in the weld but in the immediate adjacent zone.   While the weld is being made, by whatever process, this zone is unavoidably heated momentarily to a melting temperature ranging downward with increasing distance from the weld.   Such heating induces structural changes with the development of hard, brittle areas adjacent to the weld deposits, which reduce the toughness of the area and frequently cause cracking during or after cooling.

The time-temperature relations in both heating and cooling due to welding comprise, for practical purposes, a heat-treatment; and if the steel to be welded has high hardening tendencies, it may develop undesirable hardened zones in the base metal adjacent to the deposited metal. These areas are low in ductility and toughness and should be avoided, since they create a lack of uniformity in structure and properties.

An interesting study was made by French and Armstrong[6] of the increase in hardness of the base metal after welding various carbon and alloy steels.   As shown in Fig. 294, the plain carbon steel showed no marked hardening with carbon contents below about 0.25 per cent; but

with increase in carbon contents above approximately that percentage, the maximum base-section hardness increased rapidly.
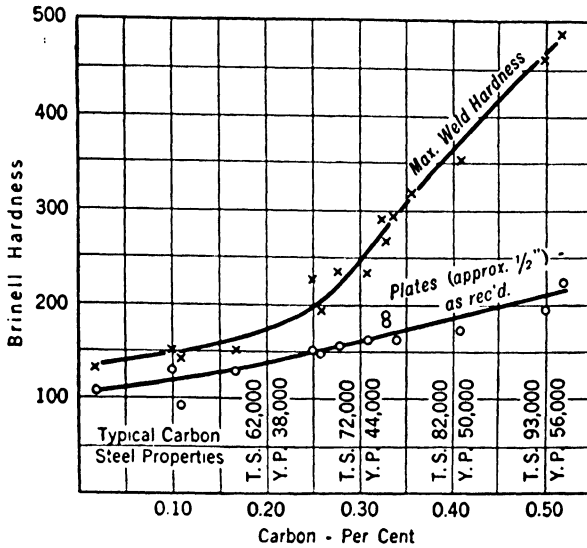
FIG. 294.—Effect of carbon content in plain carbon steels on weld hardening. (*French and Armstrong.*[6])
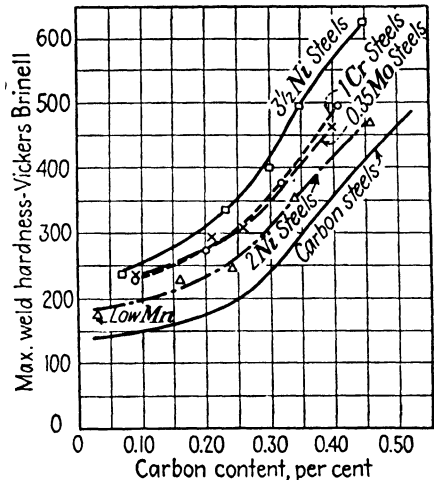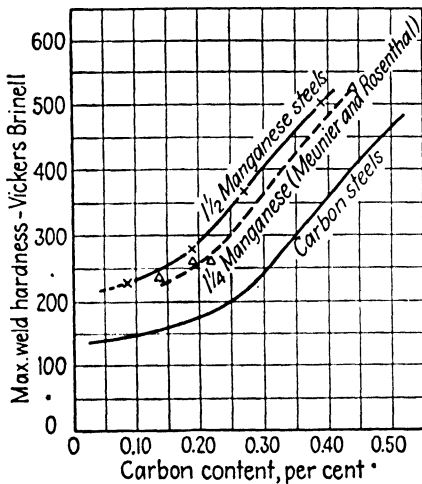
FIG. 295.—Weld hardening of steels. Maximum weld hardness in ½-in. plates of carbon steels and ternary steels of manganese, chromium, molybdenum, and nickel. Surface beads deposited at about 5 in. per min. Heavily coated $\frac{3}{16}$ in. electrodes used. *French and Armstrong.*[6])

The maximum hardness resulting from the welding of various alloy steels was also reported by French and Armstrong. Their results are shown in Fig. 295. These curves point out that manganese is the most

potent of the elements considered, next to carbon, in increasing weld hardenability.

The slope of the curves of Fig. 296, showing maximum weld hardness in relation to alloy composition, is greater at all carbon contents between 0.10 and 0.40 per cent for the elements manganese and molybdenum than it is for chromium; and it is greater for chromium than for nickel. This means that the tolerance for carbon on the basis of any allowable maximum weld hardness is least in the ternary manganese or molybdenum steels; somewhat greater in the chromium steels (to 1 per cent chromium);
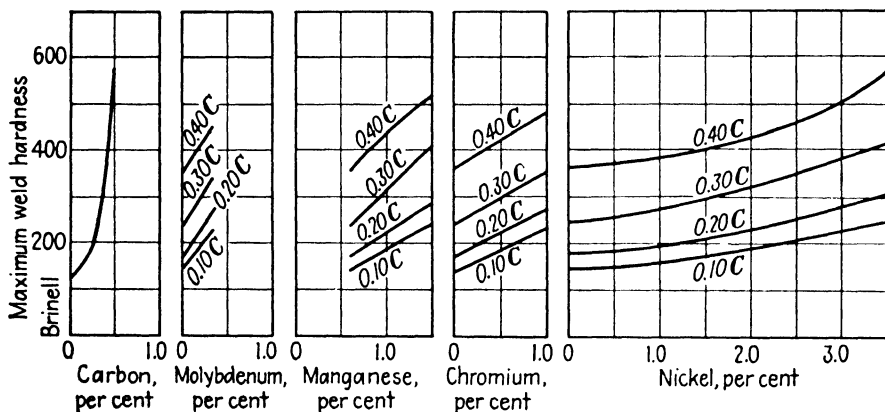
Fig. 296.—Effects of carbon, manganese, chromium, molybdenum, and nickel on weld hardenability of steels of different carbon contents. (*French and Armstrong.*[6])

and still greater in the ternary nickel steels, especially within the range up to 2 per cent.

Conclusions based on hardness results for a 0.30 per cent carbon steel led Emerson[7] to state that no particular difficulties would be encountered in the welding of this material. In most cases it may be said that the use of preheating or the use of multiple beads for the purpose of supplying additional heat to 0.30 per cent base metal is not required. If the weld is to stand high dynamic stresses, it would be advisable to use 300°F. preheat in combination with a single-pass weld, or at least a two-pass weld with the second pass immediately following the first.

A trend in some quarters to attempt to discredit the safe weldability of steels with carbon contents usually found in casting grades may have raised questions in the minds of foundrymen and casting users as to the weldability of steel castings in general. From the work of Bolton and Smith,[8] however, it would seem that such discrediting is unwarranted. Hardening propensities of the base metal as developed by microscopic examination showed that structures developed in steels of 0.25 per cent carbon were fully as good as in the steel of 0.15 per cent carbon. Steel

of 0.40 per cent carbon content required some preheating, to avoid initial martensitic formation.

According to Mueller, Smith, and Oesterle,[9] preheating is not necessary for the satisfactory welding of plain carbon steel castings having a carbon content not exceeding 0.35 per cent associated with 0.60 per cent manganese. However, subsequent stress relief is desirable. Studies at Battelle[5] showed that cast steels of varying carbon-manganese contents are in fairly consistent relation to the wrought steels. The general trend, however, was for the cast steels to show somewhat higher average maximum hardness values—probably accounted for by the higher silicon contents of the cast steels.

From the preceding discussion, it becomes evident that the weldability of carbon and alloy steels depends not only on the metallurgical characteristics of the steel being welded but also upon many variables associated with the welding technique and the design and dimensions of the parts. Therefore, tests on the base metal should be made under exactly the conditions encountered in practice, if quantitative valuation of weld-hardening effects is to be secured. With the above conditions firmly in mind it is possible to make a tentative summary of the weldability of carbon and alloy cast steels.

Carbon steel[3,4,5,9] of normal manganese content below 0.30 per cent carbon is readily weldable and requires no preheating or stress-relief heat-treatment, providing the incidental alloys are low. Carbon steel[7,10] above 0.30 per cent and below 0.40 per cent carbon is not so readily weldable and requires a preheating of 300°F. before welding, followed by stress-relief heat-treatment. The Navy maintains that, between 0.30 and 0.35 per cent carbon, the preheating temperature should be a minimum of 212°F. In the case of welding both 0.40 and 0.50 per cent carbon steel castings, a single-pass weld should probably not be used on sections of ½ in. or thicker unless a preheating temperature of 550°F. is used. If several layers are to be deposited that will gradually build up the temperature in the weld zone, then it may be necessary only to preheat to 300°F. A stress-relief heat-treatment is required for these steels.

Nickel cast steel[6,11] of below 2 per cent nickel and below 0.20 per cent carbon is readily welded and requires no preheat. Nickel steel castings of from 2 to 3 per cent nickel with 0.15 to 0.30 per cent carbon are not generally considered readily weldable, and preheat is preferable, as is a stress-relief heat-treatment. The rule is to use as small an electrode and as low a current as possible. Nickel cast steels above 3 per cent nickel or above 0.30 per cent carbon must be preheated to a minimum of 300°F. before welding and a stress-relief heat-treatment is necessary.

Cast steel containing chromium[12] between 1.5 and 6.0 per cent is air-hardening, and the carbon content should be kept as low as possible.

Castings must be preheated from 300 to 400°F. before welding, in order to prevent cracking. Castings must then be annealed or given a stress-relief heat-treatment, preferably directly after welding before the casting becomes cold. If the chromium content is less than 1 per cent and the carbon content less than 0.20 per cent, preheating is not necessary. Manganese cast steel[13] containing from 0.20 to 0.30 per cent carbon and 1.25 to 1.60 per cent manganese is air-hardening, and it is recommended that in sections greater than ½ in. the casting be preheated to 300°F. before welding. Welding should be followed by a stress-relieving treatment. Cast steels below 1 per cent manganese are classed as carbon steels. Under the A.S.T.M. welding grades a ratio of carbon-to-manganese content is in effect. In the heat-treated cast steel a composition range of 0.75 to 1.00 per cent manganese for 0.25 per cent carbon is possible, depending on the grade selected.

Molybdenum steel[14] castings containing from 0.40 to 0.80 per cent molybdenum with a maximum of 0.30 per cent carbon and 0.80 per cent manganese require preheating to 300°F. followed by stress relieving. The Navy goes a step further and requires preheating to 300°F. for castings with carbon contents over 0.25 per cent. The Navy likewise specifies a minimum preheating temperature of 212°F. for cast steel of 0.40 to 0.60 per cent molybdenum with carbon content below 0.25 per cent. It is apparent that only when the carbon content is below 0.15 per cent is a molybdenum cast steel considered readily weldable, requiring no preheating or stress-relief annealing.

There is seemingly no difficulty in the welding of low-carbon (below 0.20 per cent carbon) cast steels containing copper[15] up to approximately 2 per cent. These steels require no preheating prior to welding. Steels of carbon content above 0.20 per cent, and those above having a copper content appreciably over 2 per cent, require a preheating prior to welding. Small quantities of vanadium[16] and titanium as present in cast steels apparently have but little effect upon the weldability, and these steels are at present being treated in the same manner as carbon steels.

Silicon in the percentages used in cast steel has no pronounced effect upon the weldability of cast steel, except, as noted in the research at Battelle, that increasing silicon increases the hardenability of the steel, and somewhat higher maximum hardness values in the heat-affected zone are obtained with the higher silicon contents. Silicon contents up to 0.60 per cent are treated in the same manner as carbon steels. Finlayson[17] reports that there is no difficulty in welding low-carbon copper-silicon cast steels containing approximately 1.00 per cent silicon. Preheating is not necessary if the carbon content is below 0.20 per cent.

The effect of aluminum on the welding of steel is not entirely understood. The major importance of aluminum from the standpoint of weld-

ing is its role as deoxidizer and grain-size controller. Whether or not the steel has been killed with aluminum has little effect on the properties of the weld, according to a survey of the literature by Spraragen and Claussen.[18] Among some welders, aluminum-killed steels are considered difficult to weld. However, it is believed that they are not difficult to weld if the proper electrodes and technique are employed. Numerous quality welds in aluminum-killed cast steel are being prepared daily in the steel foundry.

The T-bend test is critical in determining ductility changes. Tests made by C. E. Jackson[1] on a cast steel treated with aluminum show that the effect of aluminum—if it could be considered as an aluminum effect— is of little consequence. The results are as follows:

Condition of Specimen	Bend Angle at Minimum Load, °
Cast and normalized	27
Cast, normalized, and as-welded	23.5
Cast, normalized, welded, and stress-relieved at 1175°F	40
Cast, normalized, and tempered at 1175°F	46
Cast, normalized, welded, and normalized	59

Composition: carbon, 0.22 per cent; manganese, 0.68; silicon, 0.36 (aluminum, 0.05 per cent added)

In the welding of commercial wrought steels, comparing silicon-killed steels with steels containing very low silicon, Jackson[2] reports:

1. Yield and ultimate tensile strengths are higher for silicon-killed steels.

2. T-bend test and bead-weld impact performance are better for low-silicon steels.

3. With mild steel and an A.S.T.M. grade EA electrode, silicon-killed steels show better bend ductility.

When weight reductions are desired and steels of higher elastic properties and strength are needed, low-carbon, complex-alloy steels offer distinct advantages over plain high-carbon steels with respect to weld hardening.[6] At equal or higher yield points such alloy steels show lower weld hardening than the plain carbon steels and have the added advantages of better retention of toughness at low temperatures and better creep resistance at elevated temperatures. The effects of alloy compositions to lower the general weldability of cast steel are additive. By that is meant that the effects of nickel and of chromium combine in a nickel-chromium steel to produce a steel less readily weldable than either of the alloying elements used alone.

Studies made at Rock Island Arsenal[19] on low-alloy cast steels show that these steels are weldable without preheating if the highest hardness

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[1] Reported by personal correspondence.

in the heat-affected zone does not exceed Vickers 465 or Rockwell A74. Yield strengths of 65,000 to 70,000 p.s.i. can be obtained with the maximum weld-bead hardness below Vickers 465 if the carbon and manganese are kept low (under 0.25 per cent carbon and 0.80 per cent manganese) and the required strength is obtained from the addition of an element besides molybdenum, such as nickel.

The preceding statements concerning the effect of carbon and other alloy additions on the weldability of steel apply equally well to a consideration of welding castings in the green or heat-treated condition. The

Fig. 297.—Section of casting showing areas from which defects have been removed. (*Underwood and Ash.*[38])

difference is merely one of stresses present in the steel in the welded condition, whereas the structures obtained from welding would be similar.

Preparation for Welding.—Steel castings are prepared for welding by first cleaning the casting surface so that it is free from attached sand and sand inclusions. The surface blemishes are removed by chipping and grinding. It is good practice to etch the scarves to see if all traces of the defective portion have been removed, especially if the defect is a crack, hot tear, or fine porosity. Defects such as shrinkage cavities, sand or slag inclusions, and large porosity can usually be followed by chipping or grinding without resorting to acid etching. It should be remembered that grinding and etching are more positive procedures in disclosing the last traces of defects (Fig. 297). Some operators prefer to use the carbon arc to see that the defective portion has been completely chipped out. The arc is passed over the scarves, causing local expansion.

Carbon dust settles in the cracks and becomes apparent. It is personally believed that the etching method is better since the carbon arc is likely to extend cracks.

Flame gouging is often used to remove defects. This is a very rapid procedure, although it is not a certain one. Operators claim that they know when they have reached the bottom of defects. In most cases they are right; but grinding and etching have shown, occasionally, that it is not too wise to rely upon the judgment of the operator of the burning torch. This is not necessary in quality welding, since the area to be repaired will always be ground or machined prior to welding.

Flame scarfing will remove defective metal more rapidly than the carbon arc, although the carbon arc is very effective in areas where the depth is greater than the extent of the surface area. Such cavities are not the best for receiving weld material under quality-welding conditions. It should be mentioned that the flame-scarfing method preheats the surrounding base metal more extensively than is usually obtained by the carbon arc. This item is important in considering the preparation of air-hardening castings for welding. In the preparation of air-hardening castings, it would be well to chip out the defect rather than to remove it by flame cutting, unless care is given to the cooling of the casting to prevent the formation of a hardened zone. Defective areas that have been flame-gouged or prepared by burning out the metal with the carbon arc should be ground before welding, to remove the oxide films and scale deposited on the scarves. This grinding need not be extensive since the scale is merely a surface effect. If the scale is permitted to remain on the scarves during welding, the electrode covering in the shielded-arc-welding method must absorb it, or oxide inclusions will appear extensively in the weld metal. The order of foundry preference in the removal of defects prior to welding is: (1) chipping, (2) grinding, (3) flame gouging, (4) machining (including drilling).

Radiographic examination and magnaflux testing are frequently used for the purpose of exploration to ascertain the removal of defects. Magnaflux testing has found greater application than radiography, since it is relatively quickly performed at any stage in the defect-removal operation, in the same manner as etching; but it is not so convenient. Radiographic examination is ordinarily carried out after welding, because it reveals not only the condition of the base metal as to defect removal but also the soundness of the weld.

The sketches in Fig. 298 show the desirable preparation of the base metal in the removal of defects, to obtain the best bonding conditions between the weld metal and the base metal. The upper sketch shows a method of preparing the base metal when chipping extends only partially through the wall of the casting. The lower sketch indicates a practice

used when, in the elimination of a defect, chipping extends entirely through the wall. In this case a temporary backing plate is used as the base of the weld to hold the first layer of weld metal. After welding, the plate is removed by chipping it off and rewelding the inside. The sketches were proposed by the Navy Department for use in the repair welding of castings.

It will be noted that if these charts are followed, considerable steel must be removed in eliminating a rather deep shrinkage cavity, if the scarves are to be prepared at the angles shown in Fig. 298. In this case

PREPARED AREA NOT EXTENDING THROUGH CASTING

PREPARED AREA EXTENDING THROUGH CASTING
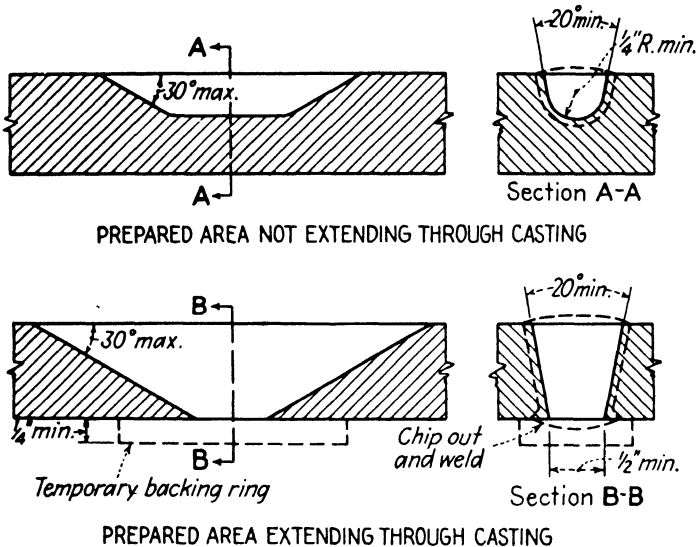
Fig. 298.—Prepared area in steel casting sections. (*Heimke.*[22])

it would be better to prepare the scarves in a U shape rather than in the recommended flat V shape.

It is believed that more care and thought should be given to the preparation for welding of minor surface defects. Even though these defects are minor, they should be carefully ground out and not be covered without some preparation of the surface. Usually sandblasting of the defective area is not sufficient, since the surface area of the defect is generally very irregular and a good bond is not established. Again, the depth of most superficial defects is greater than their surface area. This condition is contrary to that desired for the best results in welding.

Welding Equipment.—Since the welding of steel castings is a part of the manufacturing process, as careful study should be given to the purchase and maintenance of modern welding equipment as is given to any other production machinery in the foundry. A survey of welding methods used in the steel foundry indicates that the shielded-metal-arc

process is the most favored. This is in accordance with the thoughts of purchasers of steel castings, for in consumers' specifications there is a noticeable leaning toward metal arc welding. Information on the characteristics of welding equipment is available in the technical literature.[20,21]

Electrodes and Filler Rods.—Electrodes as first used were bare-wire rods mostly of Swedish iron. Studies on the effect of the arc in welding led to the use of coatings in one form and another. Out of this was developed the cellulosic type of covered electrode. The term "shielded arc welding" has been applied to welding that is performed by the use of covered electrodes of the cellulosic, mineral, or composite type. This type of electrode is used in nearly all modern-construction types of welding. It is used almost entirely in the steel casting industry. Attention is directed, however, to the fact that bare wire electrodes are still used.

One of the most obvious differences between arcs struck with bare-wire electrodes and shielded-arc electrodes is, according to Hiemke,[22] the difference in voltage for the same arc length. The voltage across an electric arc depends upon the length of the arc and the arc atmosphere, which with bare wire is practically air, although a slight contamination by the gases evolved from the melting metal may occur. In the case of shielded arcs, the arc atmosphere is a function of the electrode covering. The arc space also has ionized particles of the vapors of certain of the slag-forming constituents and metals. Because of this condition the covered electrode operates at a considerably higher voltage than bare-wire arc. From a practical standpoint, the increased arc voltage means increased arc energy, which, in turn, results in greater penetration into the base metal during welding.

Another significant dissimilarity between bare-wire welding and shielded-arc welding is in the mechanism of metal transfer. In the former a globule is formed at the end of the rod. The wire then necks down, and the globule is deposited on the base metal by gravitation or surface tension. In the case of shielded-arc deposition,

Fig. 299.—Cross-sectional sketch of a shielded-arc electrode in operation.

Larson[23] has shown that the major portion of metal is transferred from the electrode to the base metal in the form of minute droplets. Large globules of slag and metal are also deposited at intervals. The deposition of the shielded-arc electrode is illustrated in Fig. 299.

Perhaps the superiority of the shielded-arc electrode over the base-metal electrode can best be illustrated by presenting the properties of the

weld metal as deposited by these electrodes. The properties listed in Table LXXXIV are presented by Hiemke.[24]

TABLE LXXXIV.—MECHANICAL PROPERTIES OF BARE AND SHIELDED-ARC WELD METAL

Hiemke[24]

	Bare	Shielded arc
Tensile strength, p.s.i	50,000–60,000	60,000–65,000
Yield strength, p.s.i	35,000–40,000	50,000–55,000
Elongation in 2 in., per cent	6– 12	18– 35
Reduction of area, per cent	8– 20	30– 60
Charpy, ft.-lb	2	20– 40
Brinell hardness number	100–110	120–130

It is quite evident that the shielded arc develops better properties in every instance, because it is relatively free from inclusions and nitride needles (Fig. 300).
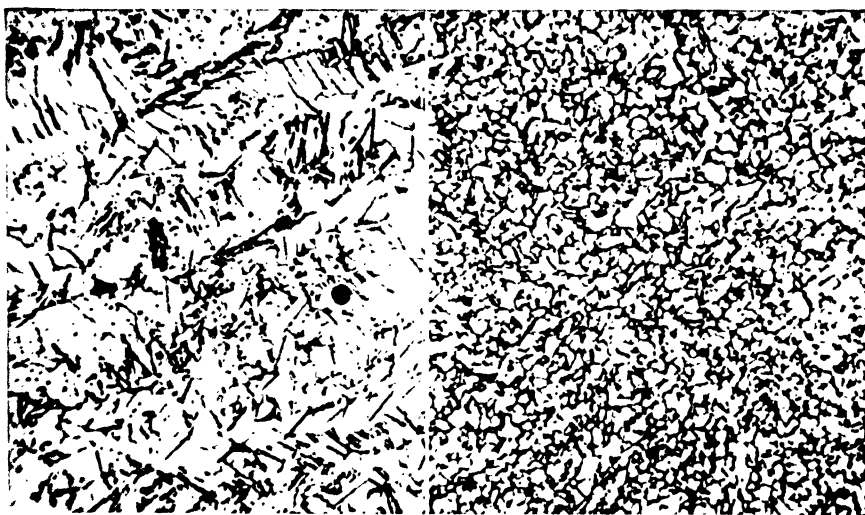
FIG. 300.—On the left, microstructure of weld using bare electrode. On the right, microstructure of weld using shielded arc. ×100. *Hiemke*.[24]

It is also claimed that the increased ductility obtained under welding conditions of the shielded arc is distinctly advantageous, particularly when welding is performed on rigid or intricately shaped structures, since the weld metal may yield plastically during contraction of the weld, which condition accompanies cooling. The material presented is sufficient to show that bare-wire electrodes have no place in the quality welding of steel castings.

The classification of the various types of shielded-arc mild-steel electrode is, according to Garriott,[25] rather difficult. They may be classified in three different ways: (1) gas-shielded, semi-slag-shielded, and slag-shielded; (2) organic, a combination of organic- and mineral-coated and a purely mineral-coated; or (3) all-position, semi-position, and position types. Basically, all the above classifications are identical except for terms of expression, and thus the construction, or the chemis-'ry, of the coverings determines the type.

The matrix of the inorganic coating consists of oxides of silicon, aluminum, manganese, and titanium in various ratios. The matrix of the organic-type coatings contains 15 to 20 per cent volatile organic substances such as wood, paper, or vegetable fibers in addition to the above-mentioned oxides.

With the organic-type electrodes, as explained by Ronay,[26] the refractory substances of the coatings have a higher melting point than the core that they surround. During the formation periods, each drop is surrounded by an extension of the refractory sleeve that holds it by capillary attraction as though it were in a small crucible. In these for-mation periods the organic substances contained in that portion of the sleeve forming the crucible become volatilized and protect the tip of the electrode from contamination by the air with a gaseous shield of a neutral atmosphere. The drop forms inside the sleeve until it develops a definite size and fluidity, whereupon the surface tension becomes insufficient to hold it within the crucible against the expelling force of the gases, generated by itself and liberated from the coating. The drop is then hurled across the arc to the base metal.

The inorganic-type electrode performs in exactly the same manner as the organic type. The difference between the two types consists in the greater volume of covering matrix per unit length in the case of the inorganic-type electrode. This greater volume enables the further prolonging of the preheating period. In some cases the coating is as much as $5/32$ in. thick.

The metal transfer is more rapid than in the case of the organic elec-trodes. The greater volume of slag transfer helps to take full advantage of the long preheat periods so that the weld metal flattens out, permitting the effortless deposition of successive layers. Inorganic electrodes are generally suitable with work only in the horizontal plane. This is no disadvantage, since their use in the foundry-repair welding is of the downhand type. The organic type of electrode is known as the "uni-versal electrode" (with $1/8$- and $5/32$-in. diameters) because, owing to the lesser volume of slag deposited, it is most suitable for welding at any angle of elevation or position.

The coverings applied to electrodes vary greatly in their chemistry.

It is not necessary, however, that a study be made of typical composi-
tions, since a more efficient way of classifying the covered electrode is to
investigate the properties of the weld metal deposited by it.   A word of
caution should be given to the effect that the type of coating must be
governed by the work to be done, for the action of the coating materials
is not the same for base metals.

This procedure has been followed by the American Welding Society
and the American Society for Testing Materials in their specifications for
iron and steel arc-welding electrodes A233.   Besides standardizing on
the core wire and its length, the specification sets up classes depending
upon the tensile strength and elongation of the deposited metal.

The classes in which steel foundry operators are mostly interested are
E7030 and E6030, since most of the welding of castings can be performed
in the flat (downhand) position.   The general-purpose, all-position elec-
trode E6010 is a recommended one and is often preferred in the foundry.
The Navy Department specification for welding electrodes describes
three classes of electrode covering.   This specification classifies all-
position electrodes as Class 1, electrodes suitable for making welds in the
flat position as well as horizontal fillets as Class 2, and electrodes suitable
for welding in the flat position only as Class 3.

Much more fluid slag and a greater volume of slag may be used with
the flat-position electrodes.   The covering of these electrodes contains
little or no combustible material.   The coverings are usually very heavy
and may be as much as 20 per cent of the weight of the core wire.   Even
heavier coatings may prove beneficial for repair welding.   Because of
the absence of cellulosic material in electrodes of this class, these may
be operated at very high current densities, which result in increased
welding speeds.   In addition to this feature of greater economy, these
electrodes produce welds of high density as measured by radiographic
examination.

The weld metal of the Class 3 (down-position) type of electrode will
produce properties of Class E6030, which, in the stress-relieved condition,
are: tensile strength, 60,000 p.s.i.; yield point, 50,000 p.s.i.; and elonga-
tion, 35 per cent.   The important qualities of this class of electrode are:
freedom from porosity, power of penetration, ease of control when
meeting obstructions, machinability, high density, and good mechanical
properties.

Shielded-arc electrodes are available for the welding of a variety of
compositions of low-alloy steels.   Alloy-steel weld metal has a higher
strength than carbon steel, and consequently it tends to have a somewhat
lower ductility.   The coverings for low-alloy steel electrodes are similar
in composition to those used on carbon steel electrodes.   In a number of
cases a core wire of the composition that it is desired to deposit is used.

In other cases the alloys are added to the covering either in the form of ferroalloys or metallic oxides. The stainless steels are also used.

It is the belief of a number of welding-electrode and filler-rod manufacturers that it is of greater importance to match the mechanical properties of the base metal than to match the chemical composition; hence in most cases these manufacturers specify the mechanical properties that may be expected from the use of the electrode rather than the composition of the weld metal deposited. There are apparently two significant reasons why the weld metal should be of approximately the same chemical composition as that of the base metal. The first is the prevention of corrosion by electrolysis in the presence of an electrolyte. The second is the avoidance of the development of thermal stresses that may assume high values at elevated temperatures because of different coefficients of expansion of the base and weld metals. The physical characteristics of the weld metal obviously must be about equal to those of the base metal.

The above discussion on matching mechanical properties in the weld and base metal is founded upon the supposition that a stress-relief heat-treatment will be given. In certain cases it is necessary that an annealing or quenching heat-treatment follow the welding operation. In any treatment involving the heating of the weld metal above the transformation of alpha to gamma iron, the weld metal will lose an advantage in strength as compared with the base metal, unless a weld metal initially stronger than the base metal is deposited.

This can best be done by going to alloy-steel or stainless-steel welding electrodes. For example, if a base metal of 70,000 p.s.i. were to be welded, an E7030 electrode would be chosen, since this electrode will produce a weld of 70,000 p.s.i. tensile strength in the stress-relief heat-treated condition. If, however, it were necessary to anneal the casting after welding instead of stress-relieving, it would be necessary to select an electrode that would produce a higher strength, since E7030 will not lay down weld metal of 70,000 p.s.i. tensile strength in the annealed condition. Since the strength of the base metal and the weld metal should be similar, a low-alloy electrode such as a carbon-molybdenum would be selected, which will give a 70,000 p.s.i. weld metal in the annealed condition. Certain foundrymen, in following this procedure, claim that they are ordering electrodes to a chemical composition rather than in accordance with mechanical properties of the deposited weld metal. Nevertheless, they are attempting to match the mechanical properties of the base metal by their selection and definitely are not attempting to match the chemicals of the base metal.

Covered electrodes are also available for welding the majority of the commercial types of corrosion- and heat-resisting alloy steel. These steels are either ferritic alloys of iron and chromium or austenitic alloys of

iron, chromium, and nickel. Since the corrosion resistance of the weld-able grades of these alloys depends upon the maintenance of a low carbon content, electrode coverings for these steels are usually free from cellu-losic or other carbonaceous materials. The grades of the various welding electrodes as produced by the various electrodes made by the different electrode manufacturers have been described rather fully in the technical literature.[24,27,28,29]

The coating of the electrode controls its polarity. Some shielded-arc electrodes are designed for best operation on straight polarity (elec-trode negative, work positive); others operate to best advantage on reversed polarity (electrode positive, work negative); still others operate well on either polarity or on alternating current. The last-named type is the flat-position welding type and is most acceptable to foundry conditions.

The filler rods for gas welding are made in various diameters from $\frac{1}{16}$ to $\frac{3}{4}$ in. These rods have manganese and silicon contents to deoxidize the weld metal. Bare-wire rods are used for the gas welding of steel. Gas welding does not permit the use of welding rods of such widely vary-ing compositions as given for electric-arc welding. Steels containing carbon over 0.06 to 0.08 per cent and manganese over 0.35 per cent develop an undue amount of gassing unless silicon is present in the rod. If silicon is present in sufficient amount, the carbon content may be increased. The commonly used welding rods are of the low carbon content and of the copper-coated type. The copper coating serves to lower the melting point of the rod and thus aids in reducing gassing. During the transfer of the weld metal a slight amount of the carbon con-tent of the core is reduced. The deposit is therefore a carbon-poor steel that has all the characteristics of such as far as tensile strength and ductility are concerned. If higher tensile strength is required of the oxyacetylene deposit, special welding rods are used. These usually con-tain a small percentage of nickel, up to about 1.5 per cent, and small percentages of vanadium. These rods are of somewhat higher carbon content, and the copper coating is not used.

Technique of Welding.—Multilayer welding, high currents, and peening should be employed for welding low-carbon castings. This technique calls for the proportioning of successive layers and using high current, which give a wider heat-affected zone and a more gradual hard-ness gradient. The current rate through the arc must be so regulated that the melting rate of the electrode will be proportional to that of the base metal. "Regulating the melting rate" of the electrode means restricting the current value to the maximum that permits the melting of the electrodes in drops. If the current rate is sufficiently high to melt the electrodes in a continuous stream, some of the fundamental requirements of properly welded joints cannot be satisfied.

Welding should be done in the flat position with $\frac{3}{16}$- to $\frac{1}{4}$-in. electrodes, the weld metal being deposited as shallow layers rather than as cylindrical beads. Peening appears to be a good way of controlling shrinkage and large grain size and should be done with a round-nose tool in an air hammer with light, rapid blows. Peening should not be applied to the first layer or two of any weld deposit or to welds in air-hardening cast steel. The weld should not be peened at a white heat, since the steel is soft and will allow the hammer to sink deep in the weld, and the grain size will not be changed greatly. If peening is done at a dull heat, the grain size will be greatly refined. Below a dull-red heat, peening cold-works the metal and, while increasing its strength, impairs the ductility.

Peening is most effective if the weld is at a dull-red heat. However, after the welder has completed a section of a weld, the weld is hot at the end but relatively cold at the starting point. If peening is to be done by the operator, he will probably cold-work the start of the weld and hot-work the hot end of the weld at too high a temperature to secure maximum grain refinement. If he can reheat the weld to the proper temperature, or if a helper follows him at the proper distance with a hammer, peening can be done at the correct temperature at a considerable expense. Peening must, of course, be confined to the weld, since peening the base metal will reduce the section thickness. Peening is used for purposes other than refinement of grain size. When properly controlled, peening prevents or corrects distortion. Peening spreads the weld and, if correctly applied, counterbalances the normal contraction due to welding. This is the most important use of peening. It may also be used to hammer down reinforcement. Blow holes, if present, may possibly be sealed by peening; but it is more expedient to correct the conditions leading to blow holes than to attempt to seal them by peening.

The arc should be kept fairly long, so that the coating does not touch the molten pool. Otherwise, porosity may result. In order to eliminate the craters at the finish of a bead, withdraw the rod slowly from the work until the arc is broken. Undercutting is caused by too rapid advancement of the arc. The solidifying metal should be watched closely and the arc advanced sufficiently slowly so that any tendency to produce defects by undercutting is eliminated.

To avoid slag inclusions, all slag should be removed before proceeding with the next bead. Scraping with a sharp instrument and thorough wire-brushing along the edges of the weld are particularly advisable. Allowing the weld to cool somewhat between beads facilitates removal of slag.

Arc blows produced by magnetic forces sometimes accompany some welding setups. An uneven burning of the electrode coating takes place and results in improper fusion, excessive undercutting, and spatter. This

condition can frequently be corrected by shifting the ground connection to another position on the casting or by changing the angle of the electrode to the work.

Low-current, small-diameter electrodes are necessary for high carbons and special-alloy castings. A fairly short arc should be used, but care should be maintained so that the coating will not dip into the molten metal. Each bead should be cleaned thoroughly before applying the next bead. Small beads should be laid down, and puddling must be avoided. In its place a weaving motion should be substituted. Metal deposited in straight beads is ordinarily not so ductile as metal deposited with the same electrodes using a weaving technique.

Stub ends of welding rods should not be used. These ends are not coated; if used, the conditions of shielded arc welding no longer exist and the process reverts to one of bare-wire welding. The foundryman will be at greater expense by using stub ends than by discarding them, because of poorer quality welds and the rechipping and rewelding that are necessary because of their use. Each foundry should give some thought as to the proper size of electrode to use for the work at hand. Certain economies can be made by proper size selection.[30]

In a recent survey it was reported that several foundries, in an endeavor to eliminate the heat-affected zone, raised the amperage and proceeded to weld with higher currents. Studies made by Pearl[31] showed that increasing the current from 90 to 100 and then to 120 amp. was not sufficient to change materially any of the properties of the weld. The heat-affected zone was found to be a little wider at the higher amperage, but the hardness and strength of this area were practically the same in all three cases.

Factors influencing the quality of oxyacetylene welding are: (1) method of welding used (backhand or forehand), (2) degree of preheat, and (3) adjustment of flame. The term "backhand" refers to the relative position of the welding rod and flame in regard to the progression of the welding. When the torch leads the direction of progress (when it is held point back to the crater), the backhand method is used. This is correct for welding ferrous metals.

The neutral flame should be maintained in the welding of steel. This flame is produced when the gas mixtures are correctly proportioned. The inner cone is larger than that used for an oxidizing flame but shorter than for the reducing flame. There is no middle cone. The outer envelope is nearly as large as that used for both the oxidizing and reducing flames, but it has no feather edge. The flame is practically soundless.

For increased ease in gas welding and for a better bond, many operators desire to preheat the casting with the torch locally in the area of the defect. In large sections a more extended zone is preheated to a

bright-red heat. The welding flame should be neutral, and filler rods flux-coated or in conjunction with a flux should be used.

Welding torches or blowpipes for oxyacetylene welding are of two general types, low- and medium-pressure. The low-pressure type operates with acetylene at pressures less than 1 p.s.i.; from 1 to 15 p.s.i. is considered medium pressure. Low-pressure blowpipes make use of what is known as the "injector principle." Oxygen passes through a small opening in the injection nozzle, producing a suction effect that draws acetylene into the oxygen stream. In medium-pressure blowpipes, oxygen and acetylene are fed at independent pressures to a mixing chamber, the construction of which varies according to the ideas of different manufacturers. Medium-pressure blowpipes, which operate with equal pressures of oxygen and acetylene, are also designated as the "balanced-pressure" type. Practically all welding blowpipes are provided with a series of interchangeable tips of different sizes.

Gas-welding practice calls for filling in the scarf by depositing as few layers as possible. Each layer of metal deposited is covered by a thin film of oxide, which must be thoroughly dissolved and slagged off during the deposition of each superimposed layer of weld metal. To avoid this complication, the scarf is usually filled in with as few layers as conditions permit. While this practice reduces the strength of the weld and increases the inclusion content of the weld, it also allows for greater heat input into the base metal because of the slow rate of progression. Because of this the weld metal and the heat-affected zone are not cooled rapidly, as in arc welding. More equiaxed grains are produced, and the weld-hardened zones are not so pronounced.

The carbon-arc process, which was used considerably in the past and not outmoded by the shielded-metal-arc developments, requires from 60 to 100 volts and 200 to 600 amp. Filler rods for this process may be as much as 0.5 in. thick. The electrode is positive.. The practice in the past has been to use uncoated rods. However, flux-covered rods give better results.

Large amounts of metal can be deposited within a short time with the carbon arc, but in so doing the methods of quality welding are ignored. The only thought is to cover the defect in the shortest time possible. Methods such as these have been responsible for the lack of regard that consumers in general have for ability of the steel casting industry to perform quality welding. This has led organizations such as the Navy to prepare process specifications that outline in detail the methods the foundries must follow in welding Navy castings. A careful study and use of modern developments in welding technique and equipment will correct this condition.

Studies have been made by the United States Navy Department and

elsewhere fairly well outlining the technique of welding from one side defects that have extended through the entire section. Welding from one side only may be successfully accomplished when the chill plate used is of the same composition as the base metal, regardless of whether the chill plate is flush with the far side of the weld or is grooved or offset. The chill plate should be about $1/4$ to $3/8$ in. thick, or about one-fourth
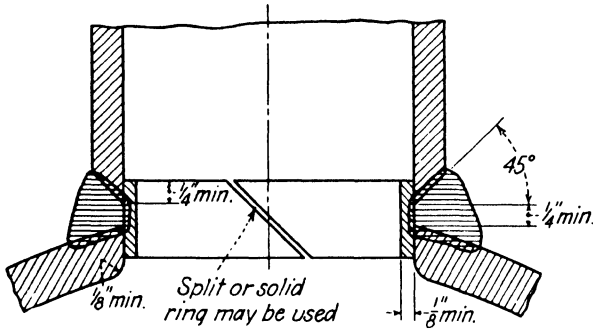
Fig. 301.—Outlet connection. Backing ring to be removed by machining after welding.

the thickness of the casting section. A sound deposit is also obtainable on the far side of the weld when the chill-plate assembly consists of a thin sheet of steel plate backed by a heavy copper plate, or by the use of a closely spaced heavy wire grill. The backing plate or ring in many cases
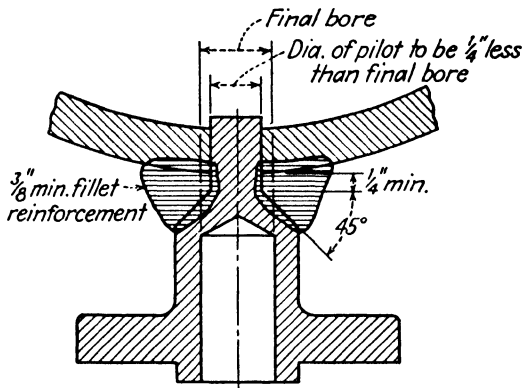
Fig. 302.—Flanged outlet connection.

is removed by machining. In some cases the backing plate may be left in place.

In the welding of a steel casting to a casting or a wrought-steel structure, the practice of using backing plates and welding from one side is a common one. A joint similar to that shown in Fig. 301 by Hiemke[22] is a very efficient one, for its use makes possible the combining of a nozzle

outlet with the main body of a pressure casting without doing any welding on the inside, at the same time retaining a completely welded structure with no unwelded margins. In this case the backing ring was removed by machining, so that it would not inhibit steam flow. Other variations of this method may be used. If the walls of the casting are heavy enough, the backing ring may be inset and does not require removal by machining; consequently, it affords no restriction to flow.

METHOD OF EDGE PREPARATION

METHOD OF FILLING IN GROOVE

PERMISSIBLE TYPES OF NOZZLE JOINTS
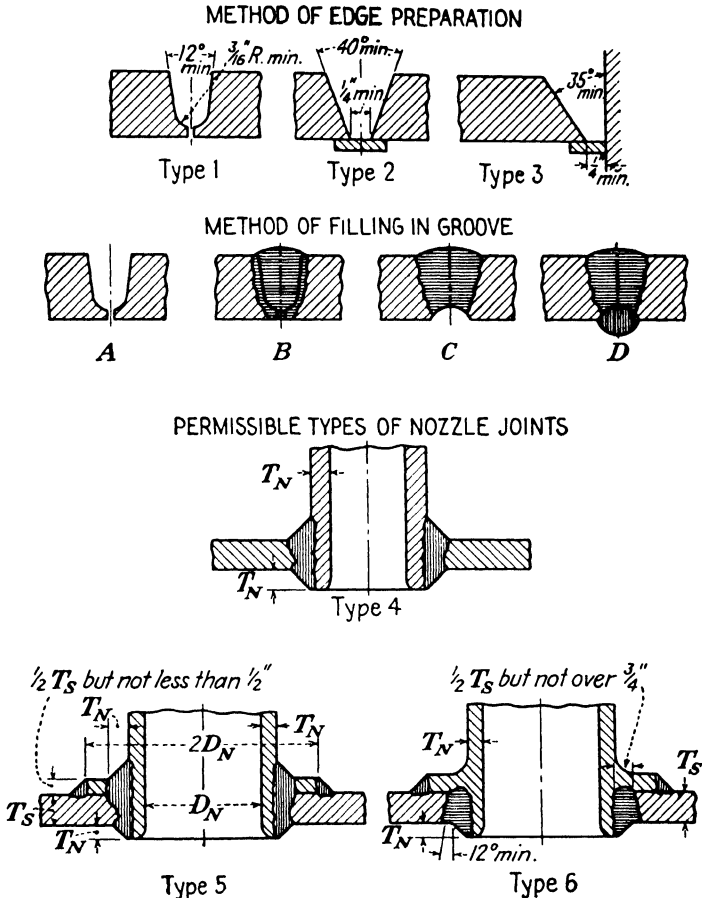
FIG. 303.—Types of welded joints. *Hiemke.*[22]

Figure 302 shows a forged or cast outlet connection of small diameter, welded to a casting by a fillet reinforcement weld. In this particular point, the solid forging or casting is used as its own backing structure for the weld. The bore is then continued by drilling, and the base of the weld is drilled out. This leaves a welded outlet with no unwelded margins subject to water or steam action. The joint designs of Fig. 303

find application in cast-weld construction or composite fabrication for pressure equipment such as turbine casings, valve manifolds, valves and and fittings, and the like.

In this particular design either a U, such as shown in type 1, or a V having a wide-open root with backing plate as shown in type 2 or the single-bevel joint shown in type 3 may be used.   In all cases, it is intended that the weld be made from the grooved side, and, after completion, the root side as shown at *C* is chipped out and rewelded as shown at *D*, thereby completing the welded section.   Types 4, 5, and 6 show methods of weld construction.   A certain amount of structural attachments such as lifting lugs and minor attachments that do not withstand high-pressure conditions may safely be attached by fillet welds.

Defects.—The chief defects found in repair welding of steel castings are insufficient removal of the defective area, oxides, and gas bubbles in the weld.   The only way in which they can be corrected is to make a close survey of the base metal by chipping, followed by etching, magnaflux, or radiography prior to the welding operation.   Besides slag and blow holes, a number of other defects are found in weld metal; these are cracks, lack of penetration, surface irregularities, pickup, nitrogen, oxygen, hydrogen, etc.

Blow Holes.—Blow holes are smooth, bright cavities of rounded contour sometimes observed in fusion weld metal.   Blow holes represent bubbles of gas in the liquid weld metal that have not risen to the surface because of lack of time and the viscosity of the weld metal.

Wormholes.—The progressive evolution of gas during freezing results in "wormholes," which are elongated blow holes trapped in the direction of the growing crystals.

Slag Inclusions.—The weld metal is almost invariably covered by a thin layer of slag, which is formed from the flux in gas welding or from the coating in arc welding.   If globules of slag are entrapped in the weld, slag inclusions result.

The following factors chiefly affect the trapping of slag inclusions:

1. Viscosity of metal.   The more viscous the metal, the more slowly the inclusions coalesce and rise to the surface.

2. Temperature.   The higher the temperature, the greater is the extent to which slag dissolves in the weld metal.

3. Rate of cooling.   The more rapid the rate of freezing, the greater the likelihood of trapping slag inclusions.

4. Stirring.   The greater the agitation of the bath, the greater the likelihood of trapping slag.

Besides the globular slag inclusions, which have comparatively little effect on the mechanical properties of a weld, elongated film-shaped slag inclusions may occur.

Film-shaped inclusions are not permissible in quality welds. They arise in two ways:

1. If the filler rod is added drop by drop and the flux is inadequate, a film of oxide forms around each drop in an oxidizing atmosphere.

2. If a new bead is started at the end of a previous bead without chipping, the scale on the first bead remains in the weld unless precautions are taken to work it to the surface.

Cracks.—Cracks that form in weld metal are similar to those that develop in steel castings. There are two types: hot tears and cold cracks. The hot tears are the result of metal-stress conditions operative while the weld metal is at high temperatures; the cracks occur shortly after the metal solidifies.

The cold cracks are mostly microscopic in size and are generally found in zones of martensitic structure; they result from stresses and the brittle character of the metal.

Cracks are nearly always parallel to the direction of welding and are usually found in the first bead to be deposited. In concave fillet beads the cracks start at the surface and are clearly visible as a zigzag or fairly straight line after welding. In convex fillet beads the cracks start at the root and may not reach the surface.

Increase in carbon content in highly killed weld metal is reported to increase, slightly, the cracking in the T-joint test. Hot peening and preheating are effective in preventing cracks in high-carbon steels.

While it is true that an increase in tensile strength of the base metal increases cracking, it is also true that there are wide differences in cracking among low-alloy steels of the same nominal tensile strength. High carbon content in the base metal favors cracking by increasing the carbon content of the weld. These root beads are particularly susceptible to cracking. Defective root fusion, attributable to too rapid welding, has been found to cause cracking.

Most investigators attribute cracking to low ductility and strength at high temperature, the weld metal being weaker and less ductile than the cooler base metal. The cracking becomes more frequent as rigidity in design increases and as the gap becomes greater. In some cases, cracks are opened by subsequent welding. If the distortion created by the later weld exceeds the capacity of the earlier weld, cracks appear. Cracks also may appear at the start of a weld, on account of distortion created by the last sections of the weld. Later formed beads of the same weld may crack because of the angular distortion aggravated by the cooling of preceding beads that have not cracked. Cracks of any type are not tolerated in quality welding, and areas that contain cracks must be removed and rewelded.

Penetration.—A defect may occur if weld metal fails to penetrate to

the bottom of any space to be welded.   Failure to penetrate in this sense is caused by a combination of too fast a welding speed and inadequate flux.   In order that the weld metal may wash up on the scarf and "wet" it, the flux must have dissolved the oxide scale caused by welding heat.

Craters.—This surface defect is the result of improper breaking of the arc.   Craters are often the source of surface cracks because the crater cools more rapidly than the remainder of the bead, and contraction stresses are set up that cause cracking.

Nitrogen.—The weld metal from bare steel electrodes generally contains about 0.1 per cent nitrogen.   The nitrogen seems to be absorbed from the air, whereas the covered electrodes prevent this absorption. Nitrogen forms nitrides in the weld metal with the resulting embrittlement of the weld.   The nitrogen content in a weld is also increased by increasing the length of the arc.

Oxygen.—High oxygen content in weld metal[32] is known to be associated with poor mechanical properties.   In bare-wire welding, a long arc promotes oxygen pickup, and values from 0.18 to 0.95 per cent have been recorded.   Values as low as 0.035 per cent oxygen have been reported for deposits from covered electrodes.   Oxygen, or rather FeO, seems to be an effective agent in reducing ductility and impact value.   Oxygen is partially responsible for aging brittleness in arc-weld metal.   There is also a belief that oxygen is responsible for the hot shortness of low-grade weld metal.

Hydrogen.—The principal source of hydrogen contained in welds[33] is in the reducing atmosphere and in moisture present during welding. Hydrogen absorbed at high temperatures develops tremendous aerostatic pressures within the steel at lower temperature, with the resultant "fish eyes" or embrittlement at discontinuities within the steel, such as at blow holes, inclusions, grain boundaries, or rifts in the submicroscopic structure.   Hydrogen can be removed from the weld by heating to temperatures below 1100°F. for periods depending upon the temperature used.   Peening the weld allows molecular hydrogen to escape through the surface rifts of slip planes that result.

Heat-treatment of Welds.   *Preheating.*—The reasons for preheating a steel casting for welding are

1. To prevent the formation of, or to reduce the hardness of, the heat-affected zone in the base metal and thereby increase the ductility and machinability of the weld area.

2. To prevent the formation of a martensitic structure in the weld and in the weld-affected area, which condition lowers the ductility and impact values of the weld area.

3. To eliminate large temperature gradients, formers of high stresses that might cause cracking in the weld.

Preheating temperatures of the base metal have been discussed previously under the section on base metal, and the requirements of various steels have been pointed out. In most cases the minimum preheating temperature was given. The casting is to be maintained at temperatures above the minimum temperature during the welding of the casting. The temperature of the casting may be ascertained by placing low-melting-temperature metal cones upon it if adequate pyrometer equipment is not available. Preheating not only affects the maximum hardness of the base metal, but it lowers the hardness of the weld metal
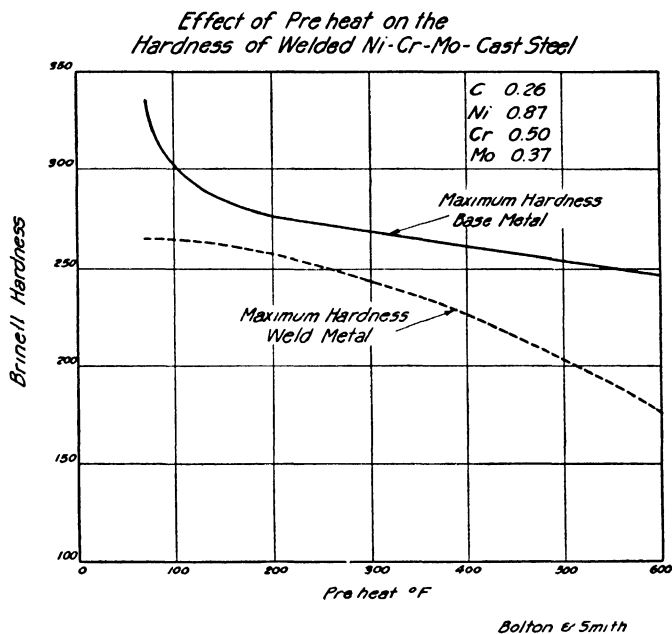
Fig. 304.- —Effect of preheat on the hardness of welded Ni-Cr-Mo cast steel. (*Bolton and Smith.*[8])

as well. The effect of various preheat temperatures on the hardness of a single-bead weld is shown in Fig. 304.

By laying additional beads on top of the single-bead weld, it is possible by the heat of welding to self-draw the hard area of the base metal. In Fig. 305 it may be noted that exceptionally high hardnesses occur in the heat-affected zone when only a single-bead weld is made. If another bead is laid on top of the single-bead weld, the maximum hardness that may be recorded in the weld is lower. A similar reduction is noticed when the third and fourth beads are laid. After the fourth bead is laid, the hardness of the heat-affected zone is not much greater than that of the base metal. It is because of this condition that multipass welds can be made on steel that has not been preheated, whereas if a

single-bead weld were to be made, it would be necessary to preheat the steel structure before welding.

It is believed that some thought should be given to the maximum temperature of preheating. The Navy Department has set a temperature of 500°F. as the maximum preheating temperature.

High preheating temperatures are responsible for grain coarsening at the junction zone of the weld metal and base metal. In a study made
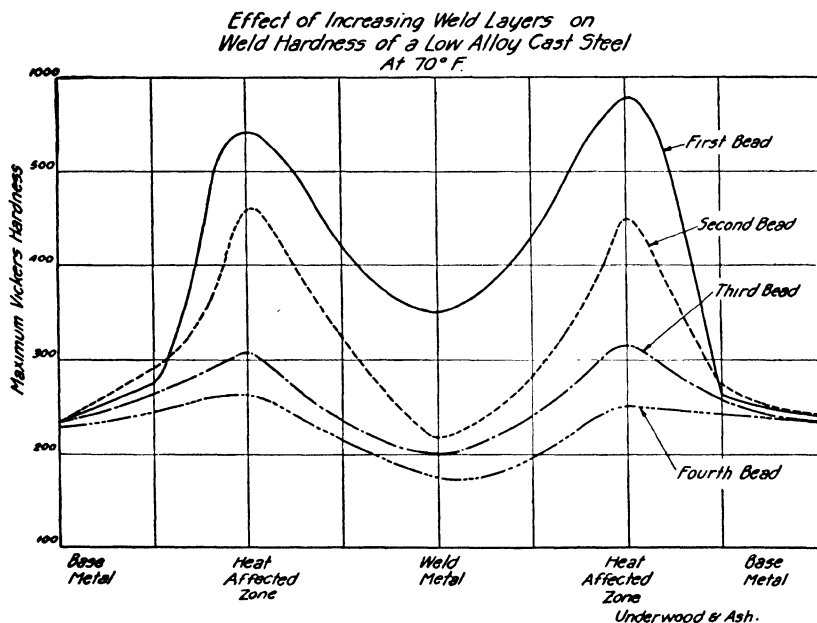
FIG. 305.—Effect of increasing weld layers on weld hardness of a low-alloy cast steel at 70°F. (*Underwood and Ash.*[34])

of the preheating temperature and its relation to the grain-coarsening tendency, Smith and Bolton[34] made the statement that none of the cast steels studied showed so marked a tendency toward grain coarsening as did the wrought steels examined. The plain carbon cast steel was most marked of the cast steels in its coarsening tendency; but even so, it was very much better in this regard than any of the wrought steels. Small amounts of alloying constituents increased the resistance to coarsening. They found that the maximum preheating temperature varied from 200°F. for the 0.35 per cent plain carbon steel to 600°F. for a low-alloy steel of the nickel-chromium-molybdenum type.

It was pointed out by Smith and Bolton that the maximum preheating temperature for the minimum grain-coarsening tendency of carbon-molybdenum cast steel (0.25 per cent carbon) is 200°F. The Navy, on

the other hand, requires that the minimum preheating temperature be 212°F. These authors claim that a preheat temperature of 300°F. would appear to be sufficient to avoid the possibility of martensitic formation in the first bead of all carbon steels (below 0.45 per cent carbon) and low-alloy steels. Moreover, the laying of a second bead would temper the hard spots in the first bead. Localized heating during the laying of the first bead would probably be sufficient to prevent martensitic formation of steels of the class 0.25 to 0.30 per cent carbon since, once started, the bead "self-draws."

TABLE LXXXV.—EFFECTS OF PREHEATING AND STRAIN RELIEVING ON WELD HARDNESS OF ALLOY STEELS WITH APPRECIABLE HARDENING CAPACITY IN ½-IN. PLATES

French and Armstrong[6]

Composition, per cent						Cast or wrought	Plate hardness, Brinell	Max. weld hardness, Vickers Brinell, when plate-welded at designated temp., °F.				
C	Mn	Si	Ni	Cr	Mo			70	250	600	900	70, then stress-relieved at 1110
0.35	1.28	0.39	1.57	....	....	Cast	230	585	415	395	...	...
0.29	0.83	0.24	2.20	0.68	0.25	Cast	255	575	480	415	...	...
0.40	0.76	0.39	2.10	0.79	....	Cast	265	610	570	455	515	...
0.44	0.45	0.24	3.50	1.48	....	Wrought	340	735	625	670	530	363
0.46	0.66	0.23	2.01	....	....	Wrought	235	470	....	....	...	300
0.23	0.64	0.18	3.27	....	....	Wrought	185	335	315	240	...	...
0.35	0.52	0.06	3.23	....	....	Wrought	205	495	370	290	...	254
0.45	0.51	0.20	3.41	....	....	Wrought	255	625	....	....	...	297
0.21	1.00	0.28	2.56	0.13	0.27	Wrought	270	475	415	330	...	...
0.35	0.70	0.26	....	0.72	0.19	Wrought	265	440	355	315	...	...
0.34	0.56	....	....	1.03	0.18	Wrought	201	502	426*	339	313	...

* Preheated at 300°F.

It has also been noted that, with increasing preheating temperatures, the depth of melting of the base metal becomes pronounced. The burning in follows the temperature of the base metal fairly closely; and, without regard for the type of material, the worst effect occurring at the highest preheats. As a result of this melting, great care must be used in welding thin-walled castings when a preheat is necessary. In preheating it is preferable that the entire casting be preheated. Since the preheating temperatures usually range from 200 to 400°F., it would be most advantageous to provide for small ovens in the welding department. This would release from this work the large heat-treating furnaces, which, incidentally, are not too well equipped to perform because of the low temperatures. Large castings are generally preheated with gas or oil torches.

Only local heating of the area to be welded is preheated. The Navy requires this local heating to extend at least 12 in. radially from the weld, and the base-metal temperature throughout this radially heated area shall be maintained at or above the required temperature.

An interesting table prepared by French and Armstrong[6] on the effects of preheating and strain relieving on weld hardness of some alloy steels with appreciable hardening capacities in ½-in. plates is presented in Table LXXXV.

Stress Relief.—It has been stated that the solidification of weld metal is accompanied by a contraction in volume that manifests itself either in movement of the base metal (warpage) or in setting up internal stresses.

Because of the rigidity of steel castings, these internal stresses around welds are likely to be rather high. Hiemke[35] states that it can be shown theoretically that the weld metal itself and the adjacent base metal will be stressed approximately to their respective yield points. The distribution of the stress in the base metal removed from the weld will depend upon the relative size of the weld with respect to the casting and the degree of fixity of the welded joint.

It is the belief of the Navy Department, as reflected by their steel casting specifications, that castings welded and not stress-relieved are likely to warp (1) while machining, because of removal of some stressed metal; (2) during hydrostatic test, when the working stresses are superimposed on the internal stresses and cause yielding; or (3) in service at elevated temperatures because of creep.

In order to relieve these stresses the Bureau of Ships requires that castings that have been welding be heated in a furnace to a temperature of 1100 to 1200°F. for 1 hr. per in. of thickness and furnace-cooled to a temperature below 500°F. This treatment presumably will reduce internal stresses to a value below 5,000 p.s.i.

In addition to its function in relieving internal stresses, this treatment acts as a toughening treatment on the weld and the base metal affected by the welding.

It is held by other authorities[3,6] that this is a very conservative position, and that carbon steels under 0.30 per cent carbon and low-alloy steels with carbon under 0.15 per cent do not require a stress-relief heat-treatment.

It occasionally happens that a casting that has successfully passed all inspection tests discloses defects during finish machining. If such castings were to be welded and stress-relieved at 1150°F., they might not be useful for the purpose intended because of probable heavy scaling of the machined surface and warpage during heat-treatment.

A steel stressed to its yield point of approximately 40,000 p.s.i. and stress-relieved at various temperatures and times would have remaining

residual stresses[35] in accordance with Fig. 272. From this graph it will be noted that temperature is far more effective than time in reducing internal stresses.

Since scale formation at 900°F. for 4 hr. is negligible and does not interfere with the proper assembly of machinery parts, it may be observed that the stress relief of a steel casting at 900°F. for 4 hr. is a practicable solution for reducing stress to such a low magnitude that any distortions taking place could be disregarded.

TABLE LXXXVI.—PHYSICAL PROPERTIES OF WELD METAL DEPOSITED WITH VARIOUS LOW-ALLOY STEEL ELECTRODES

Type of steel	Physical properties					
	As welded			Stress-relieved		
	Yield point	Ult.	Per cent elongation in 2 in.	Yield point	Ult.	Per cent elongation in 2 in.
Carbon-molybdenum 0.50	65,000	75,000	20	54,000	70,000	30
Nickel-copper (Yoloy)	60,000	80,000	20	65,000	80,000	25
Copper-nickel-molybdenum (RDS)	60,000	80,000	20	65,000	80,000	25
Copper-chromium-phosphorus (Corten)	70,000	80,000	20	65,000	78,000	22
Chrome-copper	70,000	80,000	20	65,000	78,000	22
2 per cent nickel	70,000	85,000	20	65,000	80,000	25
Cromansil	85,000	95,000	18	80,000	90,000	23
5 per cent chrome-molybdenum	75,000	85,000	*20	70,000	80,000	26

* Preheating required to obtain these properties.

The mechanical properties of weld metal at room temperature, that may be obtained in stress-relieved low-alloy steels, as compared with that obtained in the as-welded condition, are shown in Table LXXXVI.

The stress-relief heat-treatment is favored over full annealing or normalizing and tempering, since higher values are obtained in the weld metal in the former. Hardness values, of course, are lower if the full-annealing treatment is used. In some cases it is necessary that a full heat-treatment, such as annealing or quenching, be given the casting after welding. If such is the case, advantage may not be taken of the higher properties obtainable in the weld metal by stress relieving. Under these circumstances an electrode must be selected that will deposit metal of greater strength so that, after annealing, the weld metal will more nearly equal the base metal in properties.

Certain casting designs may allow the casting to be welded after a

quenching treatment. The casting is then preheated to the proper temperature for welding and welded. Further tempering and stress relieving may be done at the same time if the tempering temperature is in the region of 1100 to 1200°F. If at a lower temperature, then for a period of time the casting must be allowed to be at temperature in accordance with the results shown in Fig. 272, in order that sufficient stress relief may take place.

Metallurgical Changes Resulting from Welding.—Many welds that fail, fail not in the weld itself but in a zone immediately adjacent to it. The reason for this is that while the weld is being made, no matter by what process, this zone is unavoidably heated momentarily to a tem-
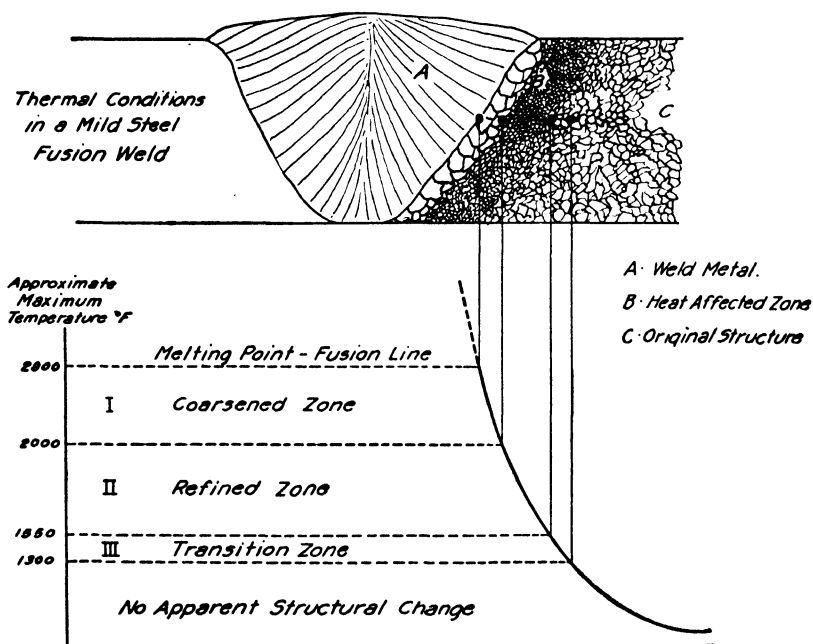
FIG. 306.—Thermal conditions in a mild-steel fusion weld. (*Aborn.*[3])

perature ranging downward continuously from the melting temperature of the steel with increasing distance from the weld. Such heating induces within this zone structural changes that at any point are determined by the maximum temperature, by the length of time at that temperature, and more particularly by the rate of subsequent temperature drop at that point; for this rate, if too great, will produce a zone of undesirable structure. If this undesirable structure persists in the welded material at the time of testing, the structure will be responsible for the failure outside the weld in the heat-affected zone of the base metal.

It can be seen, therefore, that the temperature gradient that exists

because of welding and the rate of cooling that takes place after welding are very important considerations. The base plate cools at a rate largely dependent upon the nature and size of the welded structure. If a single bead is laid with no preheating, the cooling rate will be high, particularly if the base metal is of heavy section. With the laying of more than one bead, automatic preheating takes place, and, consequently, the cooling is slower.

The temperature to which the weld metal is raised depends upon the speed of welding and power impact with the arc, or manner of manipula-
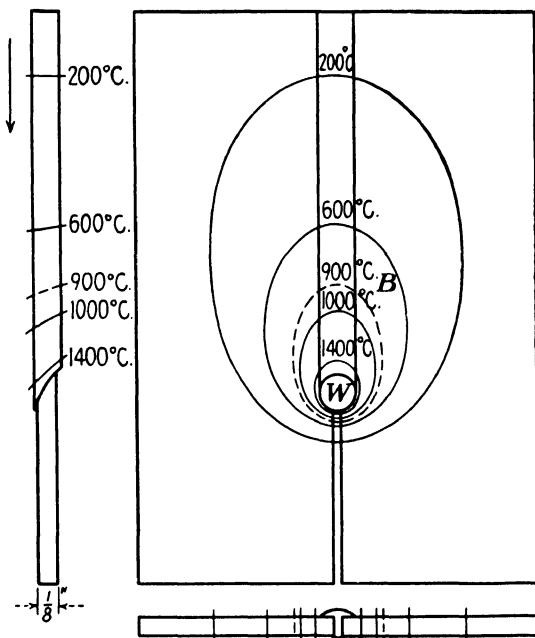
Fig. 307.—The temperature in a mild-steel plate at an instant during welding. (*Henry and Claussen.*[47])

tion if a flame is used. The heat generated at the weld spreads out by conduction into the adjacent metal and sets up in it a steep temperature gradient, from the melting temperature down to the original temperature. This gradient may be quite steep; if it is considered that the weld metal will be approximately 3000°F. and, in the case of heavy base-metal sections, metal as little as ⅛ in. from the molten metal may not be heated above the critical temperature (approximately 1300°F.), then a temperature gradient would exist at the instant of welding of upward of 12,000°F. per in. The temperature conditions and the altered zones that exist in the base metal are illustrated in Fig. 306.

The altered parent metal compresses, in three fairly distinct zones,

the several products formed on cooling: (1) a zone in which the steel was at a high enough temperature for a time sufficient to form coarse austenite grains; (2) a zone in which the austenite grains remain fine; (3) a narrow transition zone, which had never been wholly austenitic, between zones; and (2) the unchanged parent metal.

The distribution of temperature gradients that may exist in the plan view is shown in Fig. 307.

Any welded base metal that has not been heat-treated as a whole will exhibit these structural zones; they may be relatively broad or very narrow. Their width depends upon the heat input per unit length of weld and upon the rate at which this heat is dissipated. The rate of fall of temperature in and near the weld, which determines the structure and properties of the metal, depends upon the ratio of rate of heat input to rate of heat dissipation. The greater the ratio of heat input to heat dissipation, the slower the cooling of the welded joint.

The factors that change this ratio and thus influence the cooling of a weld have been stated by Aborn:[3]

Rate of heat input increased by

1. Increased energy
    a. Larger fuel consumption
2. Decreased speed

Rate of heat dissipation increased by

1. Weld-metal factors
    a. More concentrated heat input
    b. Thinner slag blanket
2. Base-metal factors
    a. Greater dimensions (thickness)
    b. Greater thermal conductivity
    c. Smaller specific heat
    d. Greater radiation and conversion
    e. Lower initial temperature

Items that can be controlled by selection are the energy, speed, thickness of the steel, and its temperature.

Let us return to the consideration of Fig. 307. The weld metal being the highest immediately after welding, it consequently has the highest cooling rate. This high cooling rate does not produce an ill effect on the properties of microstructure of the weld, since the weld is usually a relatively low carbon steel. The next fastest cooling rate is exhibited by the coarse-grain zone—the portion of the base metal that was raised to the highest temperature during welding. This combination of coarsest grains and fastest cooling makes this location the most susceptible to trouble of any region in the base metal.

If the cooling rate of this zone is slow enough, the steel in this zone will begin to transform at a high temperature, as shown in Fig. 308. If this coarse zone remains long enough within this upper transformation range, all the austenite in the base metal will transform to a relatively tough, desirable microstructure of the pearlite type. However, if the rate of cooling is so fast that the time interval within this temperature range is too short, the residual austenite then transforms (at a much lower
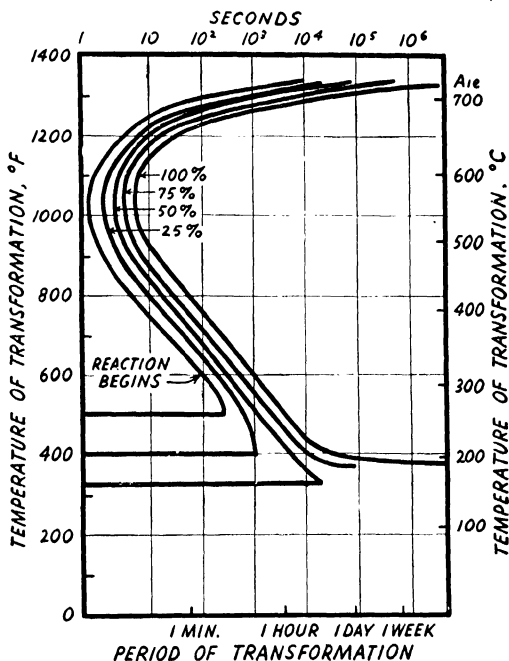
Fig. 308 —The time interval required for 25, 50, 75 and 100 per cent transformation of austenite at constant temperature in an eutectoid steel. (*Seitz—Physics of Metals*, *McGraw-Hill*.)

temperature) to a brittle, undesirable martensitic structure. The minimum time interval and the corresponding maximum permissible rate of cooling which ensure absence of martensite depend upon the composition and grain size of the austenite. The diagrams[24] of Fig. 308 show the time required for the visible onset (left curve) and the substantial completion (right curve) of the transformation, as well as the nature of the product at any temperature level.

The position of the curves on the time scale depends markedly upon the composition of the steel, which also modifies the shape of the curves. Thus, in steels such as are commonly welded, any alloying element when dissolved in austenite retards the transformation; *i.e.*, it moves the pair of

curves to the right on the time axis. Coarser austenite grains likewise displace the curves to the right.

Now, it is known that cooling takes place continuously through the transformation range. A diagram (Fig. 309) representing transformation on continuous cooling consists essentially of two separate branches, the upper zone P representing the transformation to lamellar or pearlitic structure, and the lower zone M to martensitic structure.
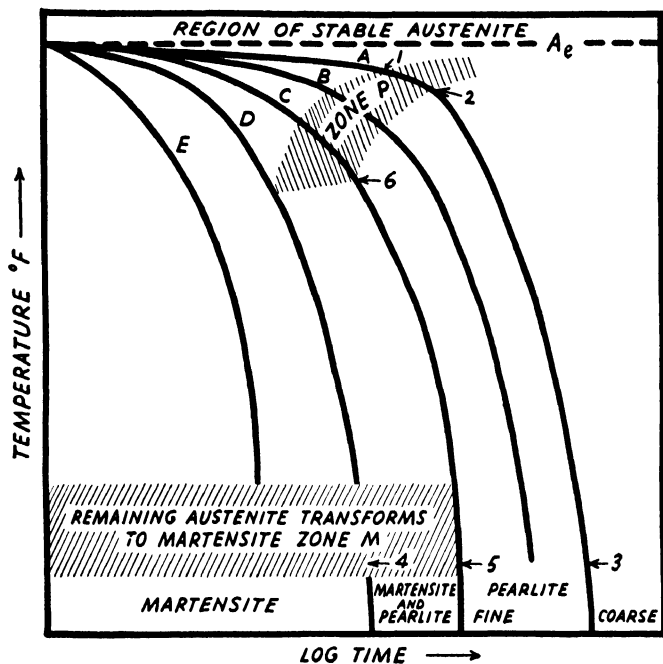
FIG. 309.—Effect of cooling rate on temperature of transformation. (*Aborn.*[3])

Several cooling curves, A to E, are plotted, each representing a constant rate of continuous cooling from the temperature below which austenite becomes unstable in steel. Actually, cooling may start from any temperature above AC, but it is only the rate of cooling below this temperature that is significant with respect to the final microstructure. In the instance of curve A, the transformation of austenite begins visibly at 1 and is substantially complete at 2. Further cooling along the curve to 3 produces no further appreciable change in structure. The product of a very slow cooling rate (A curve) is coarse pearlite. Applying the same technique to faster cooling rates, it is found that curve C represents the fastest constant cooling rate at which austenite can transform entirely to pearlite. A fine-grained pearlite is formed that is harder than that at point 3.

In specimens cooled at a rate somewhat faster than is represented by curve *C*, the austenite begins to transform to pearlite at a still lower temperature, but the time interval in zone *P* is insufficient for its complete transformation. That which has not transformed remains unchanged until it cools to zone *M*, when it transforms to the very hard, brittle, product—martensite. Specimens cooled along curve *D* just fail to form any pearlite and thus transform at the low temperature completely to martensite.

For welding, where excessive hardening is to be avoided, the maximum safe cooling rate should be slower than that represented by curve *C*.

It was stated that coarse-grained austenite moves the zone *P* to the right; thus the coarse-grained structure must be cooled at a rate shown by curve *B* in order that the structure be entirely pearlite. Preheating of the base plate retards the rate of cooling so that a normal rate of cooling such as exhibited by curves *D* or *E* would become rates as exhibited by *A*, *B*, or *C*, with the formation of the desired pearlitic structure.

Again returning to Fig. 307, why is it that there is a fine-grain zone next to the coarse-grain zone? During the heating of the base plate a transformation takes place when it reaches a certain temperature and the original structure, consisting of a mixture of ferrite and iron carbide, transforms to the solid solution of austenite. This austenite is initially fine-grained, but with increasing temperature it coarsens.

It has been shown that, because of the existing temperature gradient, certain parts of the base metal do not reach a temperature that other parts lying more adjacent to the weld reach. Thus one portion may be considerably above the transformation temperature that is conducive to the formation of large austenite grains, while in an adjacent portion the smaller austenite grains are formed because of the lower temperature. In any weld there are therefore two easily recognizable zones that temporarily were coarse- and fine-grained austenite, respectively, and these, upon the subsequent cooling, impressed certain characteristics upon the final structure. The transition zone occurs since there was neither sufficient time nor temperature to bring about a transformation.

Contraction Stresses.—Shrinkage stresses are not strictly in the province of metallurgy. Nevertheless, it is important to grasp the reason for the origin of these stresses, since the cracks and distortion are caused by them. A simple understanding of the formation of these stresses may be acquired by studying a small spot that is heated to red heat, in the middle of a large piece of sheet metal. The buckling (or distortion) of the sheet as it cools may be observed immediately upon removing the torch. During the heating the red-hot metal at the center of the sheet has attempted to expand. It has been prevented by the cooler exterior and has been hot-upset. During cooling the center contracts and the

edge metal resists the contraction. Buckles therefore form as the result of the complex stresses thus created.

The same line of reasoning can be adopted in analyzing instances of welding. Whenever large temperature gradients are established and metal is restricted in its expansion or contraction, large internal stresses will be formed. As a general rule, the stress remaining in the metal after welding is close to the yield strength. The distribution and size of contraction stresses in the upper surface of a patch weld, such as are common to steel casting repair welding, were measured by Stablein[36] using an X-ray method. A gouge 6 in. long and 1½ in. wide was filled in 24 passes, with covered electrodes depositing weld metal having a tensile strength of 74,000 p.s.i.

P= -19000
O
T= -1000

P= +23000
O
T= +6000

P=-7000   P= +33000   P= +37000   P= +23000   P,T=-27000   P= +11000
O   O   O   O   O   O
T=-4000   T=+29000   T= +26000   T= +4000   T=-13000

P= +16000
O
T= +11000

P= -29000
O
T= -16000
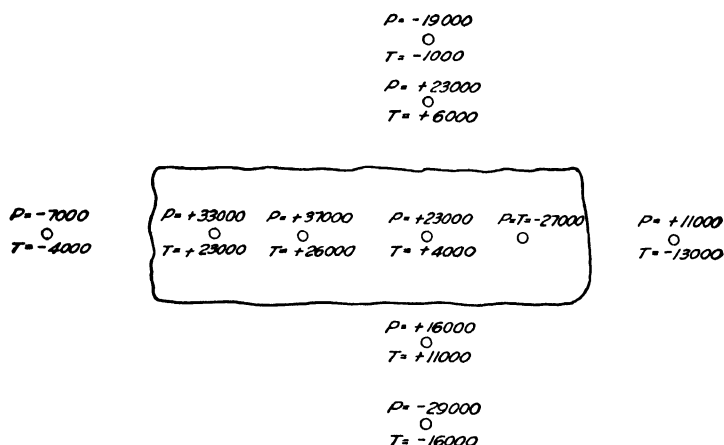
FIG. 310.—Shrinkage stresses in patch weld.   (*Stablein.*[36])

In general, the contraction stresses were tensile in and near the weld; the end of the weld, however, was under high compressive stress. (The direction of welding[37] is shown by the arrow in Fig. 310. The stresses are in pounds per square inch, P = parallel to the weld, and T = perpendicular to weld; + and − are tensile and compressive, respectively.) Stresses upward of 37,000 p.s.i. are recorded in the weld metal. It is for this reason that stress relieving is desirable.

Mechanical Properties and Welding-procedure Control.—A few mechanical properties of welds have been reported in Tables LXXXIV and LXXXVI. Further values of welded cast steel can be obtained from the literature.[7,10,22,34,38]

The mechanical properties of the weld should in all cases be similar to those of the casting. This can be accomplished by selecting the proper electrode. Since the technique of welding is also important in the attainment of the desired weld-metal properties, it is advisable that tests

be conducted to determine the quality of the welding rod and the ability of the welder. Standard methods for mechanical testing of welds have been prepared by the American Welding Society. Certain tests should be conducted regularly. These are

1. Test of electrodes for type approval and batch inspection
2. Qualification tests for welders
3. Radiographic tests
4. Process-approval tests

All electrodes used on important castings should be batch-tested to ensure their equivalence to purchase specifications, such as A233. All welders on quality work should be required to pass an annual qualification test. Each man should be required to prepare a test plate, using approved electrodes. Specimens should be cut from the test plate and bent in a standard bend-testing jig. The criterion for passing this test is soundness of the weld, particularly the bonding of the weld to the side wall of the base metal. A radiographic record of this test plate is also advisable, since it allows the welder to see what he has done in the past and comparisons can be made with the present test. All important repair welds and all cast-weld construction in pressure castings should be radiographed either by X-rays or gamma rays as an additional check on the character of the weld.

For the welding of important low-alloy steels a process-approval test should be required. This is a more searching investigation of the weld than that required by the qualification test. It has been the experience of the Navy Department that since welding of alloy steels involves a more or less special technique, including preheating and alloy-steel welding electrodes, the process-approval test is of great value.

Since the Navy Department has done more in establishing process-approval tests than any other organization, an outline of their tests as prepared by Stewart[39] is set forth for study and acceptance of such parts as seem desirable.[1]

A proposed welder's qualification test and electrode-quality test has been proposed by Walcott.[40] He suggests that for carbon steel of approximately 70,000 p.s.i. maximum tensile strength, the use of the Charpy keyhole impact test would be a good indication of quality. The electrode-quality test would result in the selection of an electrode on the basis of at least a 30-ft.-lb. impact test, the test bar to be machined from the weld deposit. Use may then be made of the quality electrode to establish a welder's qualification test by requiring a 30-ft.-lb. minimum test of the weld deposit.

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[1] Further details may be obtained from the Navy Department's General Specifications for Inspection of Material, Appendix VII, Welding: Part D, Section D-1, Qualification of Welding Process.

Cast-weld Construction.—A discussion of the subject of welding would not be complete without reference to cast-weld construction and composite construction. Intricate steel-casting design has resulted and is resulting in defective castings owing to their complicated character, which makes it impossible to allow for the necessary physical laws of solidification and proper contraction. To prevent deep-seated defects from forming, which require subsequent repair by welding, the following suggestion is offered to designers:

When the design of a cast-steel structure becomes so complicated or intricate as to cause the presence of high stresses in the casting during its manufacture, which may result in the failure of the casting, it is sug-

FIG. 311.—Steel-casting design and hindered contraction.

gested that it be broken into parts that may be cast separately and then assembled by welding or bolting.

These principles may be explained by the simple illustration of Fig. 311. This figure shows a cross section involving two features of poor design, *viz.*, the hot spot and a continuous wall section, which would be responsible for stresses mainly in the direction of the arrows, owing to hindered contraction. A casting so produced would very likely show hot tears in the central section, if the foundryman did not take precautionary steps. If, however, the designer eliminated the cross-end pieces *A* and *B*, the casting would not be under the major stresses as shown, and failure would not be a possibility. The casting could be assembled in the final form by welding previously cast or fabricated sections *A* and *B* into position.

An actual commercial application of these principles is illustrated (Fig. 312) in the manufacture of a throttle-valve casting. This sketch portrays the first design of this casting. Considerable trouble was experienced in production, and a number of attempts were made before the casting was finally produced without major defects.

In Fig. 313 the design was altered to the cast-weld assembly job. All extraneous parts were removed from the base casting and cast separately; then these parts were welded together to complete the assembly. It is needless to add that the saving in time and cost of production was considerable.

It is believed that designers should be informed that they need not be alarmed by this type of construction, for excellent properties and good homogeneity can be obtained, as has been fully demonstrated by White, Clark, and Crocker.[41]

Other difficulties that may be encountered in the production of a casting may be side-stepped by using the cast-weld construction. A
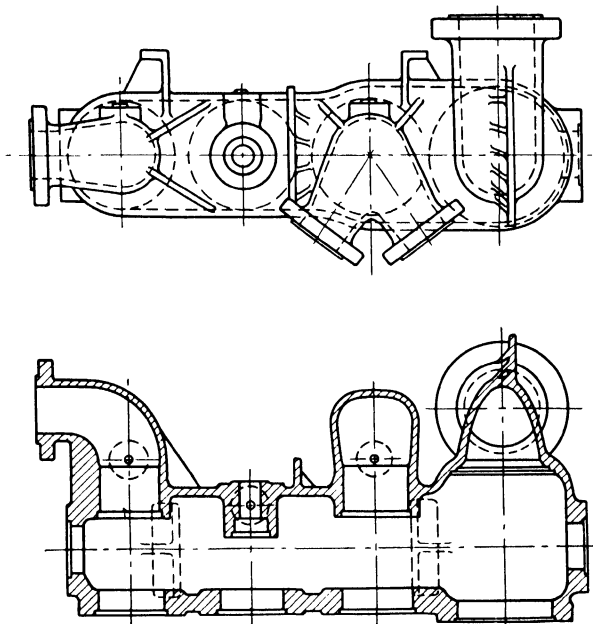
FIG. 312.—Throttle-valve casting—original design. (*Briggs, Gezelius, and Donaldson.*[48])
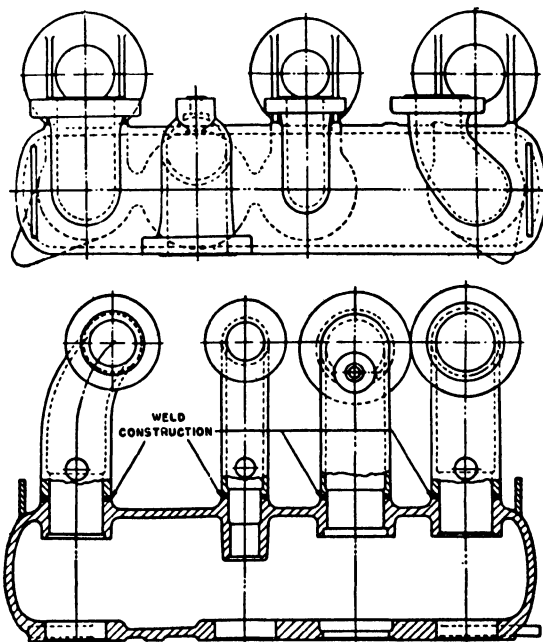
WELD
CONSTRUCTION

FIG. 313.—Throttle-valve casting—cast-weld construction. (*Briggs, Gezelius, and Donaldson.*[48])

number of illustrations covering cast-weld construction may be found in the technical literature.[42,43]

Composite Construction.—Steel castings can be, and are being, joined by welding to steel plate, forgings, stampings, and tubing, to produce finished assemblies. In a complex structure produced by welding, a steel casting can frequently be substituted for intricate welded shapes. The extent to which steel castings can be used in welded fabrication depends upon the intricacy and compactness, the quantity required, the

Fig. 314.—Gyroscope housing—composite fabrication.

direct labor, and the equipment involved in their production. The cost of a finished construction produced entirely by welding increases directly with the number of separate pieces involved and the number of inches of welding required. In the case of quantity production, the use of steel castings often has a favorable influence on costs.

Steel castings may be used in welded fabrication as an integral detail of a specific part of a complete structure, such as an intricate head casting welded to the open end of a seamless tubing. Or they may be used as an entire unit in a complete structure. An example of this latter case is shown in Fig. 314. Machinery housings are often constructed from welded plate. This is especially true when only one or two of a design are being made. An economical use of steel castings such as trunnion castings can be made in that they must be identical in shape and are rather intricate in design. Other examples are reported upon in the literature.[44,45]

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CHAPTER XVI

INSPECTION AND PROPERTIES

INSPECTION

Commercial castings today are required to meet service demands unheard of only a few years ago. The demands are increasing in severity not only because of operating conditions but because of weight-saving installations; and streamlining of design casting sections have been reduced, thereby resulting in a general lowering of safety factors.

The efforts of the technical and operating men of the casting industry are continually being devoted toward producing castings that will meet the stiffening demands and will stand up under the requirements of modern service. It is essential that castings for these services be homogeneous and of a high order of internal integrity. Nonhomogeneities that in low-stressed castings would be of little or no consequence become extremely damaging to the life of the castings, not only because they reduce the casting section but also because they lower greatly the dynamic strength of the steel, since defects and irregularities act as points of stress concentration.

Nondestructive Testing.—Examination of finished castings by a means of nondestructive testing ensures the internal integrity to the user of castings. Radiography, one form of nondestructive testing, is also extended to the checking of casting repairs. The welding of minor surface blemishes may be considered by government inspectors, if they so desire, sufficient cause for applying radiographic checking. Since most castings require some welding, even though of a minor nature, it is essential that the producers of steel castings have a full understanding of nondestructive testing if their organizations plan to make castings to exacting service requirements. Nondestructive examination of castings is becoming a part of consumers' specifications. Specifications of the Navy, Federal government, and American Society for Testing Materials contain nondestructive testing requirements. Purchasers' private specifications for steel castings are showing trends toward the adoption of nondestructive testing requirements. The oil and power companies are much interested in adopting nondestructive testing to castings they purchase.

The three commonly used methods of nondestructive testing in the steel casting industry are: (1) X-ray radiography, (2) gamma ray radiography, and (3) magnetic-particle testing.

X-ray Principles.—X-rays are a form of radiant energy, and in various respects they are similar to visible light. They travel at the same speed and obey many of the same laws as visible light (Fig. 315). Their wave length is much shorter, however, and this characteristic is responsible for their ability to penetrate materials that would absorb or reflect completely the longer waves of visible light. The wave length of light at the center of the visible spectrum is about 5.5×10^{-5} cm.; therefore, the wave length of the X-rays is only about 1/10,000 that of visible light.

The manner in which X-rays are generated is a very complex physical process. Fortunately, however, a knowledge of these principles is not necessary for the practical purpose of X-ray testing. The generating apparatus, briefly, consists of a transformer, with primary and secondary
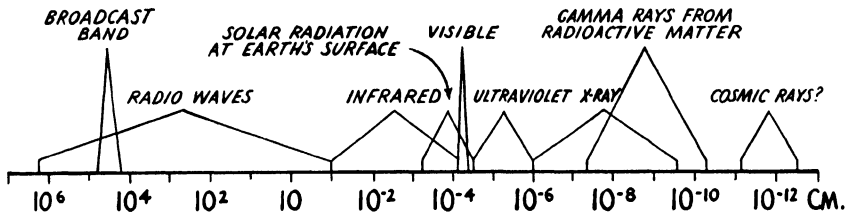
Fig. 315.—Wave lengths of electromagnetic radiations.

coils, and a rectifier. A filament transformer and the X-ray tube make up the other essential parts of the apparatus. The X-ray tube consists of an evacuated glass bulb into which are sealed two electrodes—the anode or positive electrode, which is usually a solid block of copper faced with a tungsten disk, and the cathode or negative electrode, which is a filament in the form of a coil of wire. When incandescence takes place, cathode particles, or electrons, are given off from the lighted filament and provide a bridge, as it were, between electrodes. It is possible, then, by the application of a high potential across the terminals of the X-ray tube, to drive a "pencil" of electrons from the cathode to the target with resultant production of a beam of X-rays. An industrial X-ray unit of 1 million volts is shown in Fig. 316. Principles of the radiographic process and detailed information on X-ray radiography have been carefully covered by the American Society for Testing Materials.[1]

X-ray Exposures.—X-ray testing is excellent from the standpoint of testing speed. If, for example, it is necessary to inspect a large number of small castings on a routine basis, the job could be best handled by X-ray examination. In general, exposure times are only matters of seconds or minutes, unless an attempt is made to penetrate heavy sections with low-capacity machines. An idea of exposure times for various sections of steel with X-ray machines of various voltages at a constant
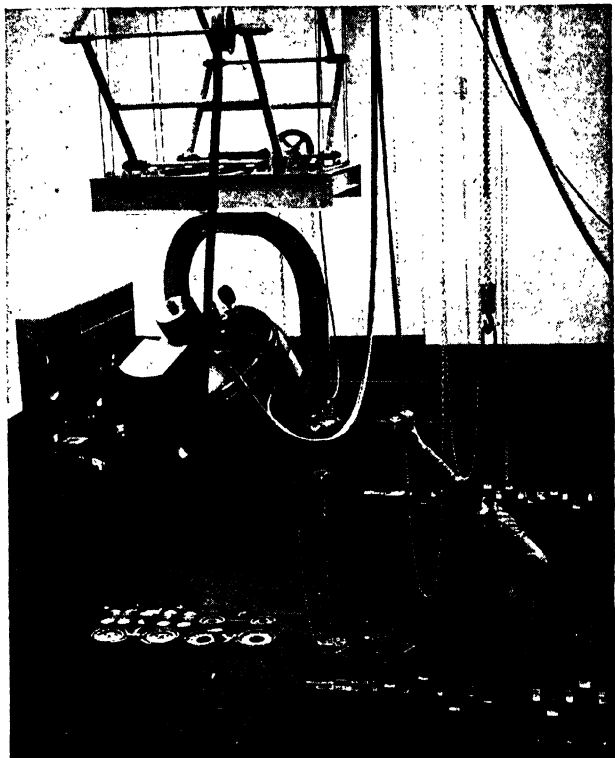
FIG. 316.—A million-volt industrial X-ray unit. Adjustment of tube head prior to radiographing steel casting. (*General Electric X-ray Corporation*)
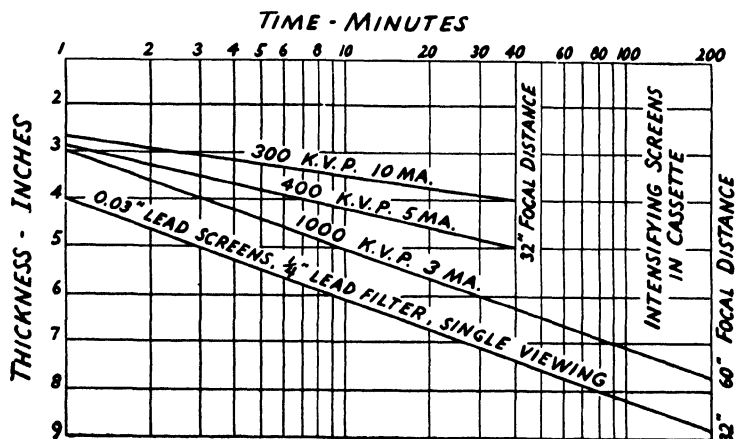
FIG. 317.—Approximate time of exposure and thickness limitations for radiographing steel with X-ray machines of increasing voltages. (*A. J. Moses*)[1]

focal is given[2] in Fig. 317. For very thin section steel castings it is not necessary that a powerful X-ray machine be used. In fact, excellent results have been obtained with machines of 100 kv. rating, although these machines are limited in their application to the steel sections. The machines that are used for the more diversified work of heavier sections are of the 200- to 400-kv. class. For large varying-section castings the 1,000-kv. machines are ideal.

The use of 100- to 400-kv. units requires operators to do considerable shielding and blocking of casting in preparation for the taking of a radiograph. An attempt is made in many cases to equalize varying sections of the casting to be radiographed by building up thin sections with lead or steel shot. Also, lead sheets are placed around the casting to absorb scattered radiation. Intensifying screens of the calcium-tungstate type are normally used in connection with radiography and the lower voltage machines. The million-volt machine requires neither blocking of castings nor the use of the calcium tungstate screens.

X-ray Advantages and Disadvantages.—The greatest advantage of X-ray radiography is the speed with which exposures may be taken.

There are several disadvantages in the use of X-rays. In the first place, it is necessary to make a capital investment of considerable proportions for equipment. Power demands and other items usually require the X-ray equipment to be permanently situated so that the product must go to the X-ray machine. The latitudes of the longer wave X-rays are not so great as might be desired. By that is meant that if a section varies from 1 to 3 in. in thickness, in a short distance as may be covered by the usual-size photographic film, the X-ray radiograph would show very little beyond the 1-in. section if this were selected for basis of exposure. This is a disadvantage, since a majority of internal defects that occur in steel castings come at adjoining sections and at these locations section thicknesses increase very quickly, and it is indeed a task to obtain a satisfactory radiography that portrays the section in detail. Sometimes it is necessary to make several shots, attempt to make a composite radiograph, or build the section up to an average thickness by blocking or embedding in metal shot or lead oxide.

Because of the much shorter wave length of X-rays generated from the million-volt machine, blocking is not necessary. The disadvantage of this machine is its high initial cost, which makes it prohibitive for use by the average small foundry.

Gamma Ray Characteristics.—Radium salt and radium emanation have been used entirely in this country for radiography. The other radioactive materials have been studied, and mesotharium has been used somewhat in Germany. However, its comparatively short life as compared with radium, with little saving in cost, is the reason that it has not been used

in industrial radiography in this country.   Radium is never used in its
pure state but in the form of one of its salts; the preference for radiography
is practically always the insoluble sulphate.   The salt is packed in a small
gastight cylindrical or spherical silver capsule.   The capsule is then placed
in a duraluminum cartridge, which can be handled by strings.

Wave-length measurements suggest a mean wave length for the gamma
ray spectrum of 0.008 A., which corresponds to a voltage of $1\frac{1}{2}$ million volts.

Radioactive elements produce three types of radiation: the alpha
rays, which are known to be helium atoms with a double positive charge;
beta rays, which are electrons; and gamma rays, which are light waves of
very short wave length.   The first product in the decomposition of
radium is radium emanation, or radon.   It is a chemically inert gas, over
100 times heavier than hydrogen.   This gas, radon, in turn decomposes
and forms successively radium A, radium B, radium C, radium D, etc.,
with the final product lead.   Each of these radioactive elements is
characterized by a certain rate of decomposition.   The strength of all
radioactive substance decreases logarithmically with time.   It has
already been stated that the half life of radium is about 1,600 years.   It
may, therefore, over a short period, be considered unchanging.   Granted
this, it may be seen that a given quantity of radium produces a constant
quantity of radium emanation each second, which in turn decomposes to
radium A, etc.

If the radium sulphate is sealed in a container, it will still yield
emanation at a definite rate.   However, if the emanation is left in con-
tact with the radium from which it forms and it builds up until equilib-
rium is reached, after which the rate of formation equals its rate of
decay, its radiating strength is then proportional to the amount of radium
present.   The amount of emanation in equilibrium with 1 g. of radium,
as measured by the intensity of the gamma rays emitted, is called a "curie";
i.e., equilibrium with 0.001 g. of radium is known as a "millicurie."   This
state of equilibrium is reached about 30 days after sealing.

The transformation of radium to radium emanation is accompanied
only by an emission cf alpha rays.   So far as gamma rays are concerned,
the radioactive properties of radium in equilibrium with its products are,
therefore, possessed completely by the radium emanation.   In the
medical practice, radium emanation is pumped from the parent radium
salt (usually in solution) and condensed from the gases with which it is
mixed by use of liquid air.   The gas (radon) is pumped into small glass
capsules ordinarily held in small brass containers.   It is quite possible to
use the radon in this form for industrial radiography.   It has been used
in the past, and some use of it is being made today; however, the use of
radium sulphate is more universal because of its longer life and more per-
manent nature, and the resulting simplicity of exposure-time estimation.

Radium radiography is excellent from the standpoint of portability. It can be moved quickly and easily. Its small size is a distinct advantage since it can be placed in cored sections and run down deep pockets of castings. Gamma ray radiographs have considerable latitude, it being quite possible to radiograph and interpret sections varying from 1 to 4 in. on one film without blocking or building sections. This is an advantage, since these varying sections are the ones that always require radiography, and a single-film record of the entire area is most desirable.
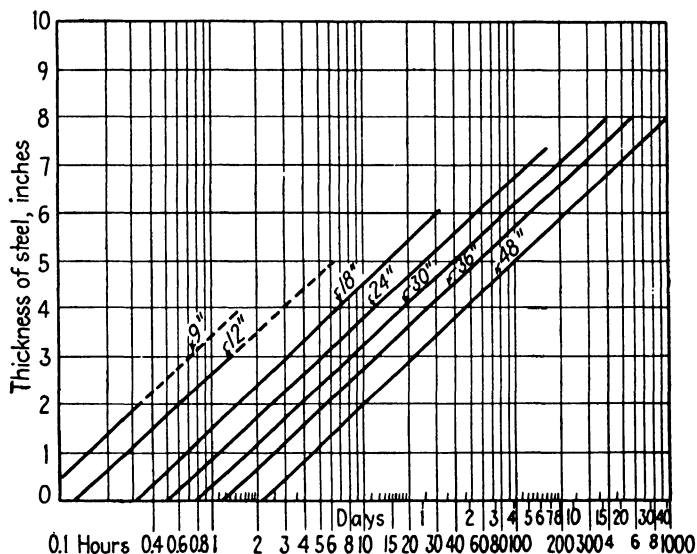
Fig. 318.—Exposure times with 100 mg. of radium. Development—Kodak, 1.6 × normal development. Distance—radium to films, as indicated. Films—Eastman Industrial Type K, viewed singly. Intensifying-screens—Lead foil. Film density—1.0 H and D. Exposure factors—Eastman; Type A–5.0, Type No-screen–2.1, Type F–1.9. (*Gezelius and Briggs.*[3])

Gamma Ray Exposures.—Gamma ray exposures are considerably longer than X-ray exposures as a rule, and hence this type of inspection does not lend itself to the routine inspection of light-sectioned castings. Its greatest use comes in the inspection of pilot castings and in the routine inspection of large- or heavy-walled castings. The exposure time for various thicknesses of steel with 100 mg. of radium at different source-to-film distances is given in Fig. 318. The exposure time varies inversely with the amount of radium used. The exposure required when 200 mg. are used would be one-half those given in the chart for 100 mg. To obtain the required exposure for any thickness of steel it is only necessary to follow a horizontal line, indicating the thickness of steel to be penetrated, across the chart until it intersects the diagonal line representing the distance from radium to film that is to be used. A vertical line

extending from this point to the lower margin of the chart indicates the necessary exposure in hours. A typical gamma ray radiographic setup is shown in Fig. 319.

When distances other than those given are to be used, the exposure necessary for the radium-to-film distance of d can be obtained from the exposure at 18 in. by multiplying the latter by $(d/18)^2$. The distance used in practice depends upon the number and size of the objects that are to be radiographed at one time and upon the definition (the sharpness of the image) that is desired. The smaller the source and the greater

FIG. 319.—Typical gamma ray radiography setup.

its distance from the defect, the greater the sharpness of the shadow of the defect. A further important advantage obtained by a greater distance is the more uniform intensity over the area of the film. It will be noted that the diagonal lines representing distances of 6, 9, and 12 in. are drawn solid up to a certain thickness and dotted to another thickness. The purpose of this is to indicate approximately the greatest thickness that may be radiographed with good results at these distances. The portions of the curves shown as solid lines can be used to indicate all defects. The portions shown as dotted lines should be used only for blow holes, sand inclusions, and larger defects, since the smaller defects such as cracks are not shown to advantage in these portions of the curves. The curve should not be extrapolated for greater thicknesses at these shorter distances if good results are desired. This fact should be kept in mind when calculating distances other than those shown on the chart. Complete instructions for gamma ray radiography and operational details have been set forth previously.[1,2]

The amount of radium used for industrial radiography varies, generally, from 100 to 500 mg. Radium may be purchased or may be rented. The rental feature is a distinct advantage, for in this way it is possible to get into the radiographic field slowly. Jobs calling for radiographic inspection may be taken only occasionally, and in this way expensive equipment is not being tied up. Over 100 foundry organizations have leased or purchased radium for industrial radiography. There are about 50 foundries participating in yearly leases.

Gamma Ray Advantages and Disadvantages.—The greatest advantages of gamma rays are their portability, small size, and the latitude of section thickness that can be radiographed and viewed at one time. Also, since gamma rays radiate in all directions, a great number of exposures may be made at a single time.

The operation of gamma ray radiography is simple; any member of the foundry staff, plant photographer, or an apprentice boy can be taught to handle the work expertly within a very short time. A trained technician such as is necessary for X-ray-equipment operation is not necessary.

The disadvantage of gamma ray radiography is that exposure times are long, as compared with X-ray exposure time. This is because the ray is so much shorter that only a minute fraction of the radiation passing through the film is absorbed by the emulsion. The shadow-image contrast is not so great as that produced by X-ray.

General Principles of Shadow Formation.—A radiograph is merely a shadow picture of an object that has been placed in the path of an X-ray or gamma ray beam. It naturally follows, therefore, that the appearance of an image thus recorded is materially influenced by the relative position of the object in the beam, and for this reason it is most important that the radiographer be thoroughly familiar with the elementary principles of shadow formation. Since X-rays and gamma rays obey the common laws of light, their shadow formation may be explained in a very simple manner in terms of light. Suppose light radiating from a point falls on a plane surface and an opaque object is interposed between the light source and the plane surface. A shadow of the object will be formed on the plane surface. The shadow naturally shows some enlargement because the object is not in contact with the plane surface; the degree of enlargement always varies according to the relative distances of the object from the plane surface and from the light source. The form of the shadow may differ according to the angle that the object makes with the incident light rays. These various deviations from the true size and shape of the object, as exhibited in its shadow image, are called "distortion." It is obvious, therefore, that in order to produce a sharp, true shadow of an object, the following conditions must be fulfilled:

1. The source of light should be small, *i.e.*, as nearly a point source as can be obtained.

2. The source of light should be as far from the object as practicable.

3. The recording surface should be as close to the object as practicable.

4. The light rays should be directed perpendicularly to the recording surface.

5. The plane of the object and the plane of the recording surface should be parallel.

Exposure Factors.—Generally speaking, the density of any radiographic image depends upon the amount of radiation reaching the sensitive emulsion of the film. The amount of radiation in turn is influenced by the distance from the source of the rays from the film, the ray-emitting power of the source, and the time of exposure of the rays. Any change in the distance between the source of radiation and the film influences the intensity of the radiation, which always varies inversely as the square of the distance. For example (Fig. 320), if it is assumed that the intensity of the X-rays emitted at the anode A in the slide remains the same, and that the rays passing through the aperture B cover an area of 4 sq. in. on reaching the recording surface C, which is 12 in. from A, then when the recording surface is moved 12 in. farther from the source to C_2 to twice the distance between A and C, the X-rays will cover 16 sq. in.—an area four times as great as that at C_1. Thus the intensity that would be adequate at C_1 must be increased four times in order to produce at C_2 a radiograph of equal density. From this information, it is possible to state the variables upon which radiographic energy depends:

Fig. 320.—Radiographic-exposure factors.[15]

X-rays:

$$\text{Radiographic energy} = \frac{\text{voltage, (kv.)} \times \text{current, (Ma.)} \times \text{time, (sec.)}}{\text{distance}^2 \text{ (in.)}}$$

Gamma rays:

$$\text{Radiographic energy} = \frac{\text{radium element (mg.)} \times \text{time (min.)}}{\text{distance}^2 \text{ (in.)}}$$

Technique of Radiography.—Detailed information on the technique of radiography has been prepared for both X-ray and gamma ray radiography.[3] There are frequently an infinite number of ways for placing

the radiographic source and films to radiograph a large object. To choose the proper technique is at times of considerable importance. The wrong choice may multiply the cost of time required or result in a poor radiograph from an observational or reading standpoint. Of course, the choice made usually involves the definition required in all radiographs, so that it is not possible to lay down rules applicable to all cases. Certain items should be remembered:

1. The sharpness of the image of a defect (definition) obtained increases linearly with the distance from source to the defect and decreases linearly with the distance from the defect to the film.

2. In general, it is sufficient to remember that an economical length to radiograph with gamma rays in one exposure is the length equal to the source-to-film distance. With X-rays only half the source-to-film distance should be used.

3. In the radiography of a flat surface, the total exposure time for the radiography of an extended plane surface is independent of the source-to-film distance. The area that can be inspected at one exposure is proportional to the square of the source-to-film distance; but the exposure time required is also equal to the square of this distance; thus the exposure per unit area does not depend on this distance.

4. In radiographing curved surfaces, the radiating source should if possible be placed at the center of the curvature.

Sensitivity.—At 8 in. of steel the sensitivity of gamma rays is 4 per cent. In the range from $2\frac{1}{2}$ to 6 in., the sensitivity is nearly constant at a value of 1.3 per cent. From $\frac{1}{2}$ to 2 in. it is 2 per cent. Below $\frac{1}{2}$ in. the sensitivity rapidly decreases to about 8 per cent at $\frac{1}{8}$ in. of steel.

In the light sections of $\frac{1}{8}$ to $\frac{1}{2}$ in., X-rays are much more sensitive than gamma rays—probably about 0.5 per cent. There is very little doubt but that X-rays can detect very fine porosity of about 1 per cent, up to 2 in. of steel. In the heavier sections, 2 to 4 in., the sensitivity of X-rays is similar in value to that of gamma rays, provided that a Bucky grid is used with the X-rays on the heavier sections in order to cut down the effect of scattering. The use of a Bucky grid is not necessary in gamma rays. Scattering may be neglected, and heavy parts may be examined with great clarity. This is an advantage in examining irregular structures and castings. The thickness of the section through which X-rays may penetrate is limited by the capacity of the machine. Radiographs taken at the upper limits are low in sensitivity, running from 5 to 10 per cent.

Contrast.—Under the best conditions, and examining a single thickness of metal of nearly uniform thickness, the contrast attainable with X-rays is greater than with gamma rays, giving sharper negatives. There are conditions, however, where all contrast may be lost because of scattering

and fogging, in which case the gamma ray method gives films showing more detail and of greater value than the X-ray method.

Safety.—It is important that the operating personnel provide and exercise adequate safeguards.   Any of the body tissues may be damaged by overexposure to the rays.   The skin, the blood, and some of the internal organs are the most sensitive.   However, there should be no cause for alarm if certain definite precautions are taken.   There are two means of protection from the rays: absorption in metal and removal to a distance.   It is convenient to use absorption of the rays in lead.   The radium should be kept in a lead safe when not in use, and the X-ray operator should be protected by a lead-lined chamber while working.   X-rays appear to be more dangerous than gamma rays because of the large amount of reflected radiation that is obtainable with this method.   It should be remembered that the reduction of the intensity of the ray is proportional to the inverse square of the distance.   Thus distance from the source is one of the best protective methods.   The X-ray operator should always remain in the shielded operating room while the radiation is on.   The gamma ray operator should place his radium and then move to a distance.   At no time is there any need for the operator to come closer than 18 in. to the radium—and that only momentarily.   As a safe-guard, technicians are required to have a blood count made at various intervals.   A person with a white count as low as 6,000 is not permitted to do further radiographic work until that count returns to above 7,000. A white count of below 4,000 is serious and is sufficient to necessitate a year's rest.   Other safeguards may be found in the use of the several obtainable instruments that record radiation exposure and its proximity.

Magnetic-particle Testing.—The principles employed in magnetic testing (magnaflux) may be explained by simple examples.   Iron filings sprinkled on a paper, supported over a bar magnet, show lines of force radiating from each pole with the greatest mass of filings concentrated near the poles.   If the magnet is broken in two, it is found that each piece has a north and south pole, and that a mass of filings has collected at the "break" (Fig. 321).   This experiment may be repeated many times, and in each instance the piece broken off will contain its north and south poles.   The pieces pushed together exhibit "consequent poles" when tested with iron filings, and the "break" in each case is indicated by a concentration of the iron filings.

If the bar magnet is bent into a circle and filings again are used to indi-cate the field, it is found that a concentration of filings appears at the two poles that have been brought near each other.   It is obvious that the conditions produced by a bar magnet bent into a circle can also be pro-duced merely by cutting a slot through a cylinder magnetized circularly, by passing a current through a wire at its center (Fig. 322).   In the solid

cylinder the flux flows continuously, and there is no appreciable external field or noticeable poles. Once this continuous circular flux path is broken by a slot, there are two poles and flux leakage—conditions similar to those created by bending a bar magnet into a circle. Similarly, if a break is caused in the circular flux lines of a circularly magnetized solid
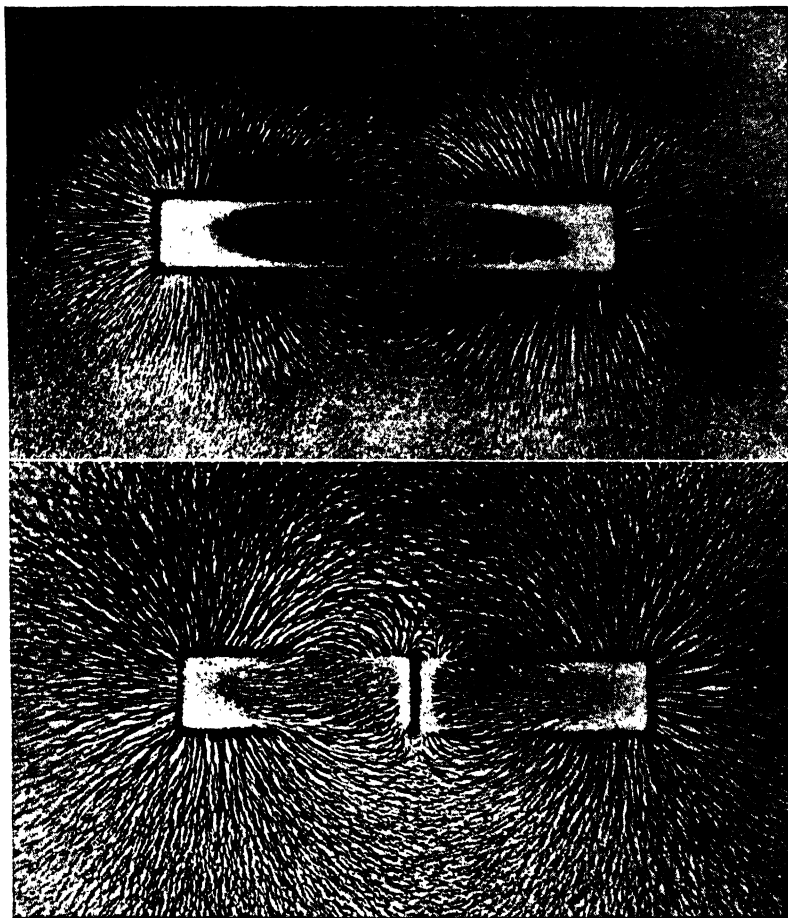
FIG. 321.—Top: Lines of force radiating from a bar magnet. Bottom: Lines of force radiating from a break in the bar magnet. (*Gezelius.*[5])

bar, local poles or local flux leakage is created that can be indicated by the use of iron filings, as in the simple experiment with the bar magnet. These local poles or local flux leakages occur wherever there is an abrupt change in the magnetic permeability of the test piece. Thus several slots in the circularly magnetized bar produce a series of consequent poles, each of which can be detected by the same means.

The illustrations given have been produced by shaking iron filings on
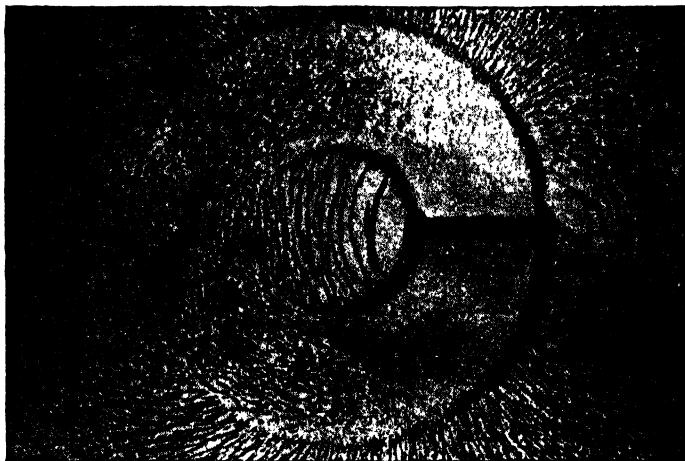
FIG. 322.—A slot in a magnetized steel tube.   (*Gezelius.*[5])
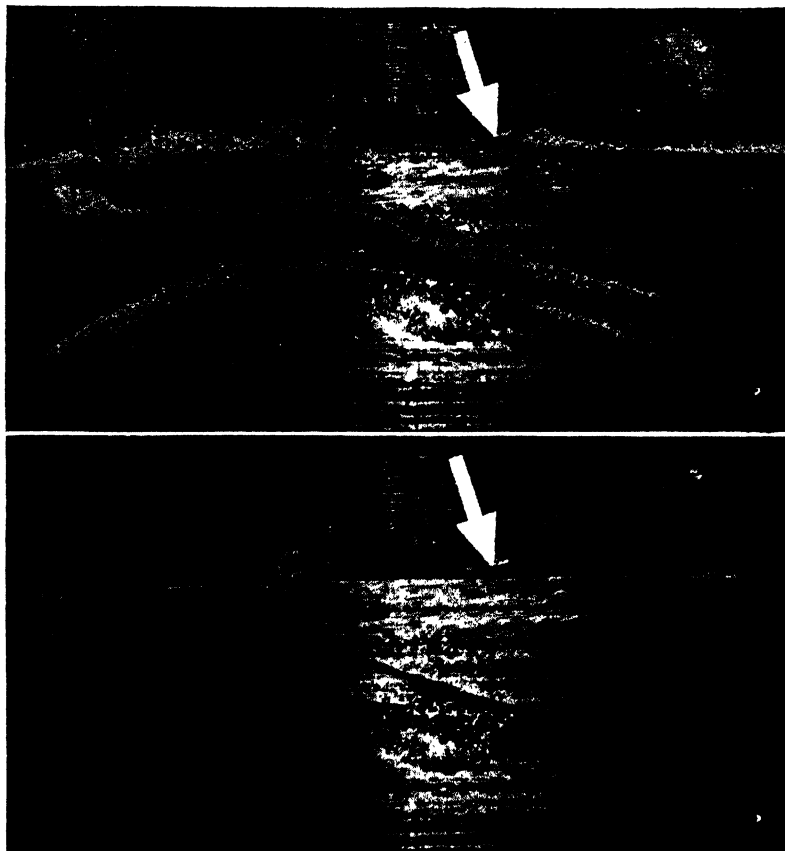
FIG. 323.—Top: Iron particles outlining cracks; arrow pointing to crack which is partly subsurface.   Bottom: Rough machined surface—visual cracks.

a piece of white paper, placed over the magnetized piece. Figure 323 was obtained by using iron particles to bring out a series of crack defects. It is interesting to note that an additional portion of the crack is revealed by using the powder.

It is not essential that an actual slot or crack produce these local poles. They may be caused by the presence of gas or shrinkage cavities; by the presence of monmetallics such as sand, slag, etc.; or by abrupt change in the permeability of the material caused by variations in heat-treatment, high stresses, etc. To summarize, two basic conditions are required in the magnetic-particle type of magnetic testing:

1. An abrupt change in the permeability within the path of magnetic flux flowing through a magnetized piece of material that produces local poles or local flux leakage.

2. A method of detecting these local poles. This condition is satisfied by the use of finely divided ferromagnetic powders, which offer a lower reluctance path than a liquid or air and hence tend to gather in the leakage field and define the outline of the field.

This method of inspection is applicable only to magnetic materials. Alloys that are only faintly magnetic do not produce satisfactory results. Defects in material tested by this method produce magnetic discontinuities. Magnetic-particle testing outlines these discontinuities. The outline produced by the powder is termed an "indication." It should not be assumed that all indications are the results of defects. The shape of the magnetic discontinuity is most important. Best results are obtained when the magnetic discontinuity cuts the lines of force at 90 deg. Long, narrow discontinuities parallel to the lines of force give very poor indications, or none at all. It is evident that magnetic discontinuities near the surface are readily detected, whereas those lying below the surface have diffused leakage fields resulting in diffused outlines that are difficult to interpret. Porosity can be detected when it is near the surface and in deeper locations where the cavities are large and few in number. If deep-seated porosity is composed of a large number of small holes, the possibility of detection is slight. In all cases the outline of the magnetic discontinuities and the depth at which they can be found are dependent upon the strength of the leakage field present; this is dependent upon the shape of the piece under test and the strength of the magnetizing force.

Methods of Magnetizing.—Magnetic tests may be conducted by using residual magnetism in a previously magnetized piece, or by the so-called "continuous" method wherein the powder is applied while the test piece is under the influence of the magnetizing force. The latter method is preferred for the low- and medium-carbon steels, and only this method will be discussed here.

The type of magnetization desired depends upon the material to be

tested and the location to be tested.   The following remarks are there-
fore general and do not apply to all conditions.   Details on the technique
of magnetic-particle inspection have been prepared.[4,5,6]   The three
methods found in common practice are (1) the "all-over" method with
the entire casting as the conductor, (2) prods, and (3) solenoids or coils.
For the all-over method, copper bus bars are clamped to opposite sides
of the casting.   This method is suitable for use in connection with weld-
ing generators.   Prods consist of copper bars attached to wire leads.   The
prods are placed on the casting at about 8- to 16-in. intervals, depending
upon section thickness.   Prods may be used with either a welding
generator or storage batteries.   In the solenoid method a conductor is
wrapped around the test piece; a current passing through the coil mag-
netizes the piece by induction.   Direct current, 300 to 700 amp. at low
voltage, is usually used; but alternating current is suitable for some types
of test.   Alternating current may also be used with proprietary equip-
ment that converts it to broken alternating current or pulsating direct
current.

Inspection Medium.—The powder to be employed should have the
following characteristics:

1. High permeability
2. Low coercive force
3. Low retentivity
4. Good mobility

After properly magnetizing the part, the powder may be applied by
either of two methods, commonly called "dry" and "wet" methods.
For either method, proprietary powders of the proper size and shape are
available that have colored coatings to increase the contrast with the
background and thus facilitate testing.   Other powders not of a pro-
prietary nature have also been used and give satisfactory results.   The
powders that are used for the dry method are usually treated to increase
their mobility and prevent agglomeration of the powder.   The powder
may be applied from a hand shaker, bulb blower, or mechanical blower.
The use of the hand shaker is limited to nearly horizontal surfaces,
whereas the blowers may be used on vertical or overhead surfaces.   In
the wet method the powder particles are suspended in liquids such as
kerosene, gasolene, or carbon tetrachloride.   A fire hazard is sometimes
involved, which requires the use of carbon tetrachloride; and if it is used,
ventilation must be provided, since the fumes are poisonous.   The liquid
may be sprayed or flowed over the part to be inspected; if it is flowed,
care should be exercised so that potential indications are not washed
away by the force of the stream.

Surface Preparation.—The surface is important in this type of testing.
In general, the smoother the surface, the better are the results.   In

examining rough castings, careful interpretation of the results is essential, for many minor surface imperfections may appear to be serious defects. All heavy layers of rust, scale, and grease should be removed prior to testing. Thin, even layers of rust or scale do not cause much difficulty.

Demagnetization.—Demagnetization is an important consideration in some applications. Many methods may be employed, and the subject should be studied thoroughly if demagnetization is required. The simplest method that is applicable in many installations is a large a-c solenoid through which the pieces may be passed and slowly withdrawn. Demagnetization is not required at all if

1. The piece is to be heat-treated at a temperature high enough to cause demagnetization.

2. The piece has low retentivity. (This includes all low- and medium-carbon steels and some low-alloy steels.)

3. The piece is to be magnetized in final assembly.

As previously pointed out, all magnetic indications do not represent material defects. Therefore, results should be clearly interpreted in order to be of value. The approximate location, number, and size of indications should be reported and described as fully as possible. Cracks, porosity, inclusions, etc., should also be identified. This information, together with the service requirements of the part, must be considered when the acceptability or repair of a part is in question.

The magnetic-particle method cannot compare with the radiographic methods. This method is useful primarily to detect surface defects and defects that come very close to the surface. The method is used mostly in routine examination for locating cracks caused by machining, grinding, quenching, or pickling. A permanent record, as is obtained by the radiographic methods, is not obtained in magnaflux inspection. It should be said, however, that fine, hairline surface cracks are detected by magnaflux, whereas these cracks may not be detected by radiography. Some of the castings inspected by magnetic-particle testing include locomotive crossheads, draft-gear housings, couplers, crankshafts, gears, valve bodies, rolls, brake drums, gun mounts, and miscellaneous tank castings.

Nondestructive Testing Standards.—Nondestructive inspection requirements in contracts, orders, or specifications should be viewed with caution by the foundry management, and acceptance should not be made of unqualified orders calling for castings subject to nondestructive testing. By so doing, the right is given to casting purchasers to radiograph or magnaflux any and all parts of castings and reject the castings even though the defects are present in unimportant sections, such as tie-in sections that bear no important stresses and may be designed so that their factor of safety is extremely high. Purchasers of castings should also safeguard themselves by not using general clauses calling for nondestructive

testing, since manufacturers will increase the cost of the casting to compensate for the undue hazards of producing castings to undetailed inspection requirements.

To protect both the foundryman and the purchaser from unqualified nondestructive inspection requirements, the foundry management should request that the purchaser submit marked drawings designating the places that will be subjected to nondestructive examination. Such drawing will help materially in the planning of production of the job. It will enable the operating department to provide for desirable metal feeding and guard against the use of metal inserts; in fact, the mold may be produced in a manner differing entirely from normal production procedure.

This request for designation of positions or radiographic examination is not an exceptional one and should not be considered unfavorably by the customer. If the customer does not apparently know where examination should be conducted, the foundryman should help him to crystallize his thoughts by suggesting that his designing engineer be contacted regarding the important stress centers.

The foundry management should also at this time discuss with the consumer the subject of interpretation of the radiographic results. In other words, the foundryman should know whether or not the consumer has a set of standards; and if so, he should discuss these standards with the customer to ascertain if they are reasonable and mutually satisfactory. If the customer has no standards, the foundryman should suggest either that a set of nationally known standards be adopted or that rejection be determined by a joint council of consumer and foundryman. The foundryman could at this time also take up with the consumer certain features in the design of the proposed casting that may be detrimental to the production of a homogeneous casting. It should be remembered that the customer will reject a defective casting as shown by radiography, even though he has designed it so that it could be nothing more than a defective casting.

It is a wise plan to have radiographic examination accomplished at the location of the casting manufacture. Such examination should go forward as soon as possible after the casting of the first unit of the lot, so that disclosed defects can be corrected in subsequent castings.

The foundryman should recognize that there is a need for two types of standard: (1) steel castings for high-pressure, high-temperature installations; and (2) steel castings for structural purposes. For example, a greater degree of porosity could be permitted in structural castings than for the other type. Lack of fusion as exhibited by internal chills may not be considered objectionable in structural castings, whereas they may be considered so in high-pressure, high-temperature castings. In other words, it does not appear that the degree of homogeneity in pressure cast-

ings would be necessary in structural castings, and thus at least two sets of standards could be devised.

Certain consumers, notably in the U.S. Navy, have prepared radiographic standards that are used in the rejection or acceptance of castings. The U.S. Navy Bureau of Ships standards for steel castings consist of gamma ray and X-ray radiographs classified as to (1) serious defects that require the casting to be rejected, (2) defects that because of their localized character may be repaired, or (3) minor defects that are acceptable without repair. These films are used as standards of comparison with those submitted under routine examination.

It appears that standards for the magnetic-particle type of non-destructive testing are rather difficult to develop. In view of this situation it is suggested that inspectors be carefully trained in the importance of the magnetic-particle indications. A standard testing procedure should be developed and used.

Destructive Testing.—Occasionally, a buyer of castings specifies that a casting or several castings, depending upon the size of the order, are to be tested to destruction. These castings are placed on an anvil block and are broken by dropping a heavy metal ball from a sufficient height. The sections are then examined. The casting tends to break at points where the section is vulnerable to the applied dynamic load. The break may not necessarily occur through a defect, since the elements of casting design are such that stresses applied in the test are different from those contemplated by the designer. The presumption that the casting will break at defective places is not necessarily correct. In some cases the metal saw is used rather than the drop ball, to ascertain soundness. This procedure, although it is a slower method of testing, allows the casting to be sectioned systematically and deep-acid-etched to reveal major and minor weaknesses. Torch cutting of castings may also be used, but only major defects are encountered.

A test of increasing importance is one wherein the casting is placed in a large tension-testing machine and tested in tension or compression until failure is recorded. Such testing gives the design engineer useful information on the acceptability of the design for load requirements found in service. A great deal more of such testing should be done, since it is believed that through such testing more confidence can be placed in steel castings. The only objectionable feature is that testing is limited to small castings because, except for a few installations, machine capacities are not sufficient to do the testing required.

Inspection requiring the destruction test is best suited to large-quantity production where molding methods, gates, risers, and pouring methods may be standardized. Destruction tests may be performed on the first casting from the pattern and faulty procedure on design cor-

rected. Subsequent destruction tests are of value to the purchaser, for they indicate the internal conditions of the castings as a lot, all castings having been manufactured in the same manner.

MECHANICAL PROPERTIES

The mechanical properties of carbon and alloy cast steels have been completely and systematically reported in the "Steel Castings Handbook,"[8] and only a brief summary will be presented in this section.

The properties of annealed carbon cast steels are given in Fig. 324. The composition of this grade other than carbon is manganese (0.50 to 1.00 per cent), silicon (0.20 to 0.70 per cent), phosphorus (0.05 per
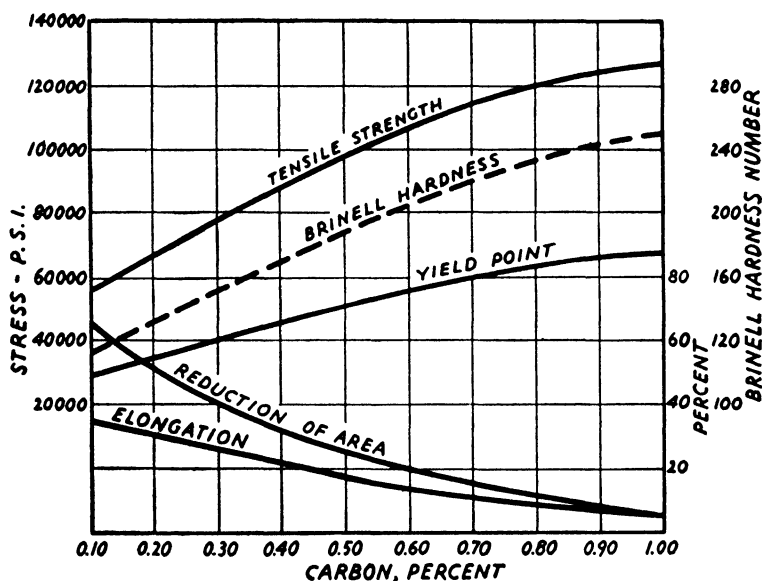
Fig. 324.—The mechanical properties of full-annealed carbon cast steel.

cent maximum), and sulphur (0.06 per cent maximum). The greatest bulk of the steel casting production falls in the carbon range of 0.20 to 0.50 per cent. The most popular heat-treatment given is normalizing, although normalizing followed by tempering is in considerable demand. In Fig. 325 the mechanical properties of the carbon range of 0.20 to 0.50 per cent are shown in detail for various heat-treatments. Figures 324 and 325 are compiled from those values that are normally expected from commercial cast steel and are obtained from distribution curves plotted from testing results of separately cast or attached 1-in.-square section coupons.

Studies by Wellauer[9] on wrought steel, cast steel, and weld metal showed that if a definite hardenability or hardness is the basis of comparison, the tensile strength of rolled, wrought, and cast steels and weld

metal are identical, regardless of the alloy content; the minimum yield values are somewhat proportional, but naturally they are more scattered. Thus for design purposes involving tensile and yield properties, the engineer may interchange rolled, wrought, cast, and welded steels with the
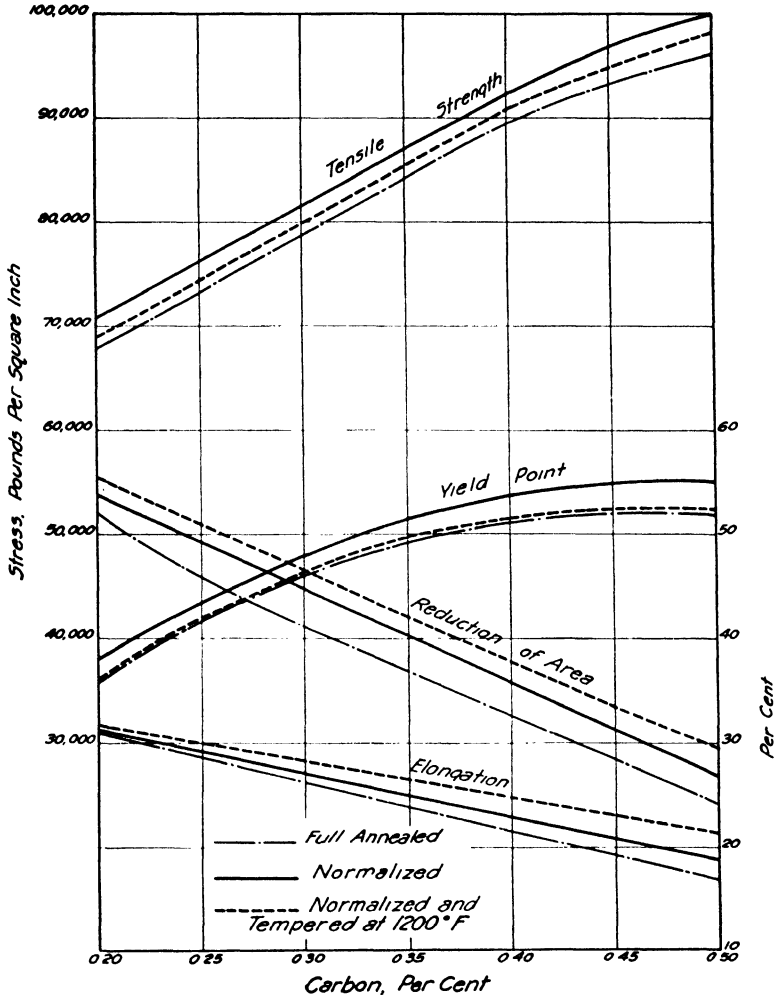
Fig. 325.—Comparison of full-annealed, normalized and tempered properties of medium-carbon cast steels. (*Steel Castings Handbook.*[71])

fullest confidence. Also, Wellauer showed that the ductility properties of the various forms of steel construction are nearly identical. The longitudinal ductility properties of forged products are slightly higher than those for cast steel or weld metal. However, depending upon the degree of working, the transverse properties are much lower. Since most service conditions involve several directions of loading, the security of

uniform directional properties of cast steel is sometimes particularly advantageous.
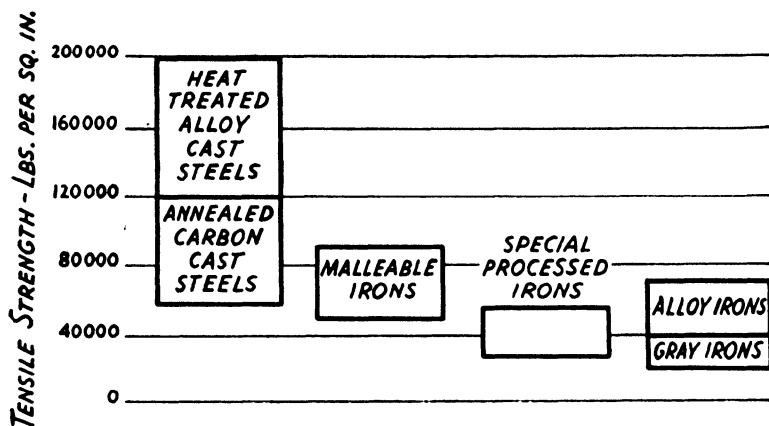
FIG. 326.—Ranges of tensile strength of cast ferrous metals.[10]

The manner in which cast steel compares with other cast ferrous metals as to tensile strength and impact resistance is given in Figs. 326 and 327, for a 0.13 to 0.80 per cent carbon cast steel and for various alloy
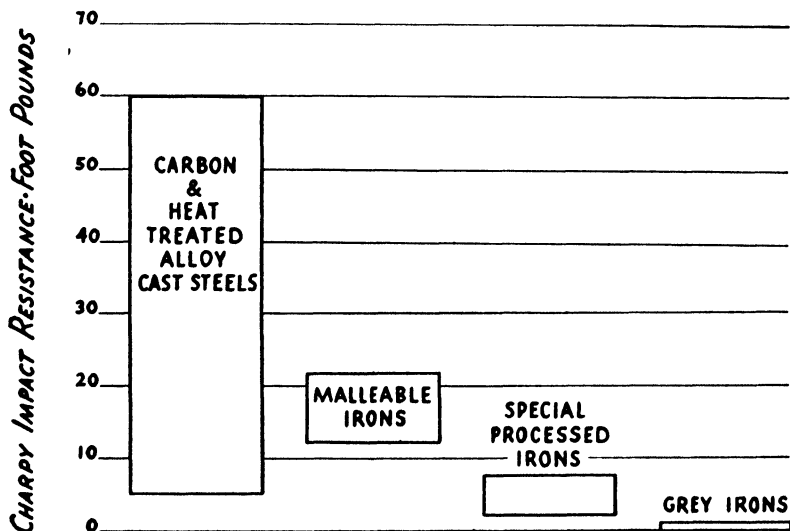
FIG. 327.—Ranges of impact resistance of cast ferrous metals.[10]

cast steels.   In Fig. 328 a comparison is made of the reduction of area values of all cast ferrous metals.

The heat-treatment that a cast steel receives has considerable effect upon the mechanical properties of the steel, as shown in Fig. 325.   This

effect is more pronounced when quenching and tempering treatments are used, as is illustrated in Fig. 329 where the yield-point property of a 0.30 per cent carbon steel is studied. Note particularly that the yield point is increased by all types of heat-treatment and can be more than doubled
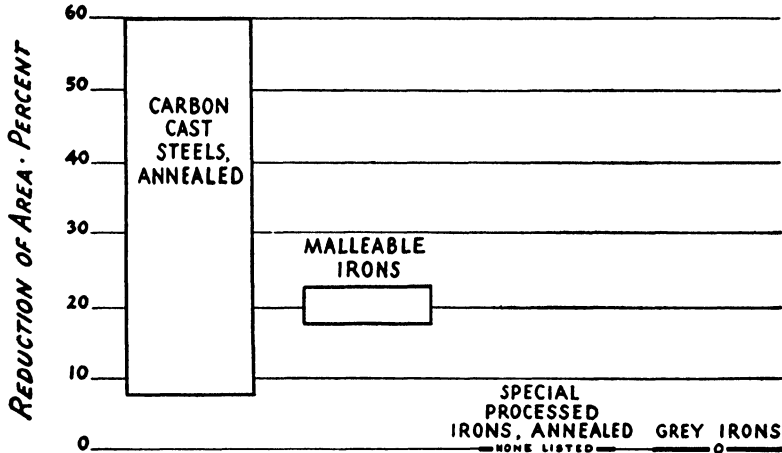
Fig. 328.—Reduction of area for cast ferrous metals.
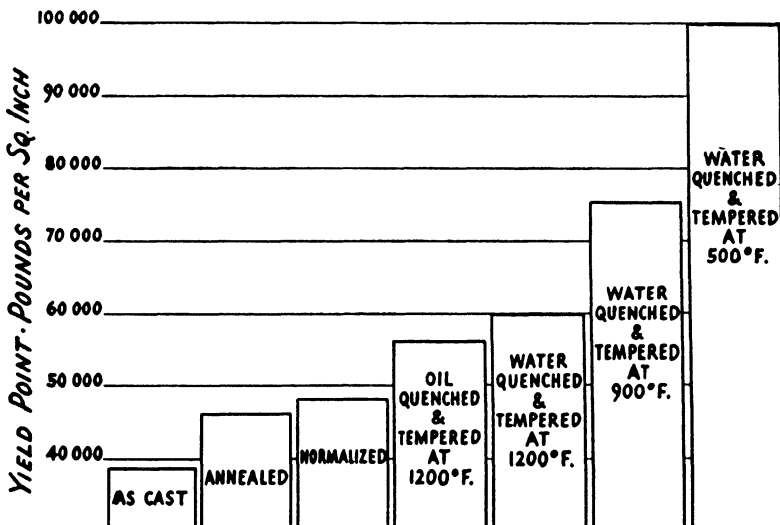
Fig. 329.—Yield points of a 0.30 per cent carbon cast steel after different heat treatments.[10]

according to the heat-treatment applied. The temperature at which cast steel is tempered following a quenching treatment affects both strength and ductility properties. This is illustrated in Fig. 330.

The response of alloy cast steels to heat-treatment is at least as impressive as in the case of plain carbon steels. For example, Fig. 331

shows the response to heat-treatment in the case of a nickel-chromium-molybdenum low-alloy cast steel.   Each type of heat-treatment imparts
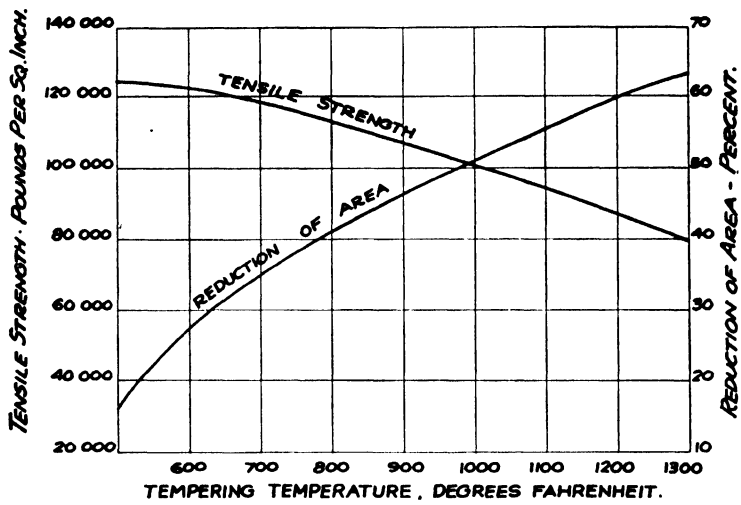
FIG. 330.—The effects on tensile strength and reduction of area obtained by varying the temperature at which 0.30 per cent carbon steel is tempered after quenching.[10]
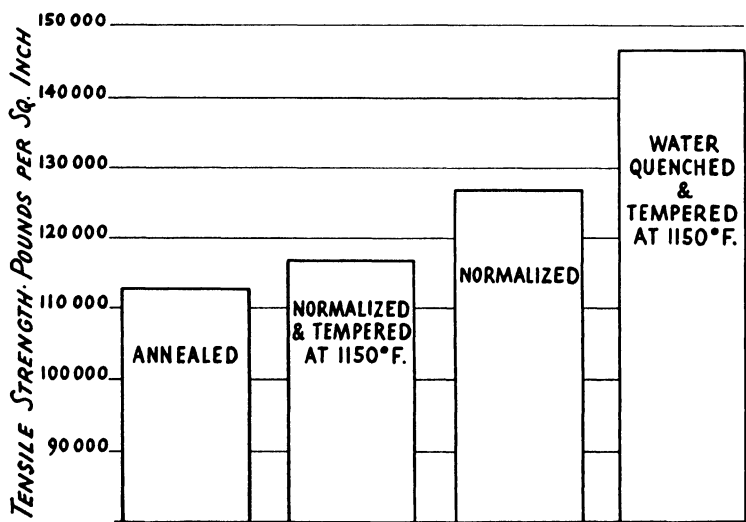
FIG. 331.—Tensile strength of an alloy cast steel (carbon 0.30 per cent, nickel 1.68 per cent, chromium 0.55 per cent, and molybdenum 0.27 per cent) resulting from different heat-treatments.[10]

special properties to the steel.   Moreover, it is not always necessary to use an elaborate or expensive heat-treatment to secure the required mechanical properties.   The normalizing treatment that is especially

designed for certain of the low-alloy cast steels may be all that is necessary to obtain adequate properties. It is noteworthy that, in quench and tempering, a wide range of properties is possible by the simple device of
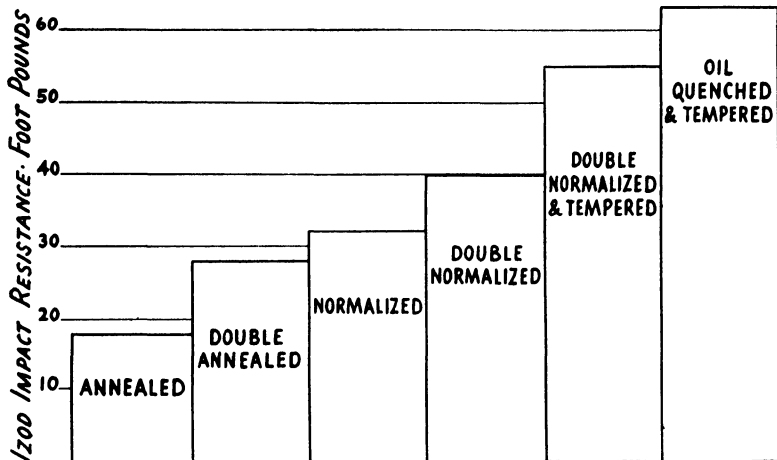
FIG. 332.—Impact resistance of an alloy cast steel (carbon 0.30 per cent, nickel 1.50 per cent, and vanadium 0.14 per cent) resulting from different heat-treatments.
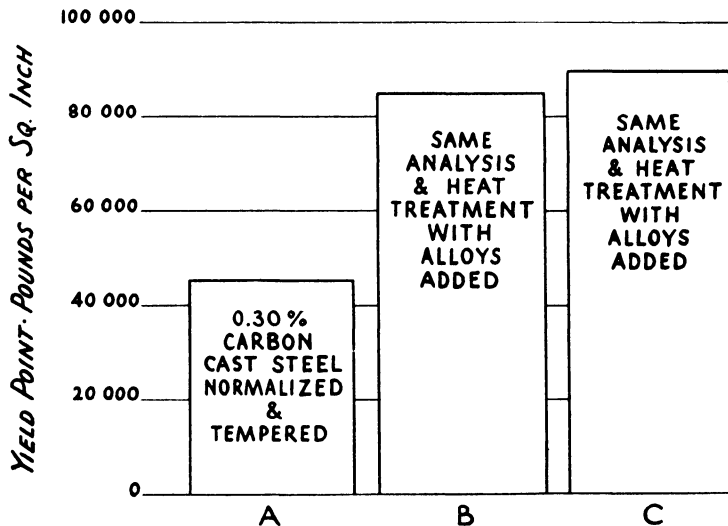
FIG. 333.—Yield points of a 0.30 per cent carbon cast steel resulting from the addition of alloys (in the case of *B*, chromium 2.00 per cent and molybdenum 0.50 per cent have been added; in the case of *C*, nickel 1.20 per cent, chromium 0.70 per cent, manganese 1.55 per cent, and molybdenum 0.30 per cent have been added.[10])

varying the tempering temperature. Heat-treatment has a definite effect upon the impact resistance of alloy cast steels such as the nickel-vanadium steel shown in Fig. 332. The quench and temper treatment

increases the impact resistance by more than 200 per cent. Figure 333 shows the effect of alloy additions on the yield point of a 0.30 per cent carbon cast steel. By the addition of alloys, it is possible to increase the yield point as much as 100 per cent.

Brinell Hardness Relations.—The linear relation of Brinell hardness and tensile strength is well known. An estimate of the tensile strength may be made by multiplying the Brinell number by a factor of 500. It was found by Wellauer[9] that the relation BHN $\times$ 500 fitted very well for a large number of commercial carbon and alloy cast steels, of both
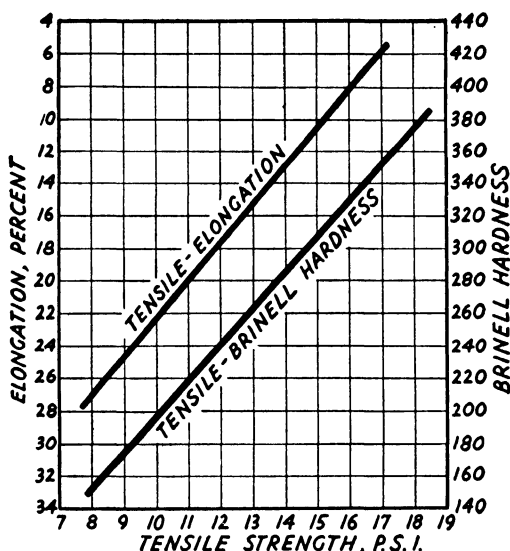
Fig. 334.—Relation of tensile strength to Brinell hardness and elongation of a single steel in various stages of heat-treatment. (*Schwartz and Bock.*[12])

normalized, and quenched and tempered heat-treatments. This same relationship was studied by Ruffle[11] for 100 consecutive heats of normalized medium-manganese cast steels. A relation of ultimate strength = BHN $\times$ 450 was an average value recorded. The tensile properties of medium-carbon low-alloy cast steels were studied by Schwartz and and Bock.[12] The relation obtained for 25 cast steels is shown in Fig. 334. The data are applicable only to steels that have been normalized and/or quenched. The steels, if normalized, may or may not have been tempered. If quenched, they must have been tempered above 750°F. Also as shown by Fig. 334, if the tensile strength for each heat-treatment is plotted against the corresponding elongation, the locus is invariably very nearly a straight line. Other relations between Brinell hardness and yield stress have been recorded for various cast steels.[11,12]

The composition of steel can be used to calculate its hardenability,[12]

but a relation of hardenability to any mechanical property of cast steel is by no means sufficiently accurate, quantitatively, to permit any close predictions as to what values of any properties may be expected to accompany given hardenabilities.[14]

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INDEX